METALS & ALLOYS

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The Magazine of Metallurgical Engineering

INCLUDING

CURRENT METALLURGICAL ABSTRACTS

SHEET NORMALIZING FURNACE AT OTIS STEEL COMPANY (*Courtesy of Surface Combustion Company*)

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METALS & ALLOYS

The Magazine of Metallurgical Engineering

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CURRENT METALLURGICAL ABSTRACTS

EDITORIAL COMMENT

The A. S. T. M. Meeting

THE METALLURGICAL sessions at the A.S.T.M. meeting late in June dealt with much of interest and value for reading and study, but as technical sessions they were, on the whole, tiresome. The Symposium on Steel Castings was an exception. It went off well because there were no committee reports presented for action at those sessions.

The high spot of the Symposium was the presentation by John Howe Hall of his paper on Austenitic Manganese Steel Castings. No one was bored by that talk and everyone will remember it with pleasure. He is added to our Hall of Fame.

The Round Table on Acquisition of Good Data came so early in the week that a poor attendance was inevitable. Some of the papers at the Round Table were sensible and useful and deserve detailed study.

On the whole the other sessions were dragging the anchor of ponderous committee mechanism and made the writer feel that since everything passes through letter ballot anyhow, that part of the convention dealing with adoption of specifications might far better be conducted on the correspondence school basis, with opportunity offered for objections that need, perhaps once in a decade, to come on the floor of a general meeting. The present mechanism is both complicated and rusty and it creaked a lot in 1932.

Had that part of the program been eliminated, there would have been time for the technical papers and for discussion. Attempts were made to present in abstract papers that had not been preprinted and were entirely unsuitable for brief presentation. In some such cases the authors got away with a difficult task very well, but it was generally a waste of time, since it brought out no discussion. Oral discussion appeared to be a lost art, but only because no one wanted to start anything and make the sessions drag out still longer.

There must, of course, be some mechanism to safeguard against the railroading through of a specification, but it should be possible to develop a satisfactory method that will not waste as much time as the present one.

Until a better and more modern mechanism can be installed, the specification end of the job ought to be separated from the research end, at the annual meeting.

It is often said that the real work of the Society on specifications is done behind the scenes by the committees and subcommittees. We'd like to see them kept behind the scenes still more instead of throwing a deadly pall over what would otherwise be lively and interesting sessions.

We repeat that an audience has some rights.—H. W. GILLETT.

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Metallurgical German

BACK in September, 1929, we commented on the metallurgists' need for a sufficient working knowledge of technical German to be able to follow German work in the original. Engineers and scientists can find so much of value in German publications that to ignore them leads to wasteful duplication of effort. Merely to follow them through abstracts is not enough. When one is chasing a subject down in detail, getting full translations made of all pertinent German articles is expensive. The metallurgist needs to be able to handle German himself if he is to be effective and up-to-date. We pointed out in the former editorial that college courses planned with this aim in view would be useful, but that they might well differ from the usual course in "literary" German.

Experience along this line at Battelle Memorial Institute may be of some interest.

About 30 of the staff decided that they needed to brush up their German and a course of 20 lessons was arranged, under the leadership of Professor C. W. Foulk of Ohio State University. Geitz's Metallurgie was selected as textbook. Some of the students had had several years of German, others but one year, and two had had none at all.

Professor Foulk is professor of Chemistry, not of German, though he has travelled and studied extensively in Germany, and his methods were quite different from those of the average professor of German. He recommended that those who had had no German study grammar concurrently with his course, but for those who had some knowledge of grammar, only a slight brushing up was suggested with the idea that the grammatical requirements could, to a considerable degree, be painlessly absorbed much as one absorbs them when learning a language as a child.

He laid great stress on the vocabulary, telling us that if we knew what all the words meant we would almost automatically know what the sentence meant, and we were drilled and drilled on vocabulary. He would not stand for a "smooth" translation, but required that we translate word for word in the German order, thus making sure that the meaning of every word was understood, and preparing us for the time when we could sense the German by reading it in German without translating. At the end of those 20 evenings even the two who started at scratch had a working knowledge of metallurgical German and those who had had some introduction beforehand estimated that their ability to handle it had increased from two to six-fold. The average student probably spent about 4 hrs. per lesson in preparation. The total expenditure of 100 hrs. time was certainly well repaid.

This experience has convinced us that any group seriously desiring to get a reading knowledge of technical German, which can find a leader with anything approaching the ability of Prof. Foulk, can make very real progress if they go at it with equal earnestness. If such effort is to be taken up by metallurgical groups in given plants or localities, we would caution them to avoid professors of "literary" German, but get a scientist who knows German to lead the group. The metallurgical faculty in a University is handicapped in putting anything of this sort across because the faculty of German is prone to rise up on its hind legs unless it is allowed to instruct the metallurgists according to its standard methods of instruction. But we estimate that to get equivalent training by way of the "literary" courses would take at least 4 times as long and be far less satisfactory.—H. W. GILLETT

Light Alloys in Motor Vehicle Construction*

By FRANK JARDINE†

WHEN considering the past history of pleasure cars, buses and trucks, it is realized that performance has received more attention than any other feature. A new model is hardly in production before new ways to increase the top speed, the acceleration, and the economy are being considered. There has also been a decided tendency to increase the size of all vehicles whether they are trucks, buses, or pleasure cars.

The demand for better performance and larger vehicles usually results in heavier vehicles requiring more powerful motors. In some cases it has been found necessary to use sixteen-cylinder motors in pleasure cars and twelve-cylinder motors in buses to obtain satisfactory performance. These large motors have made the problem of weight distribution rather difficult and a chassis built to carry a large powerful motor at high speeds becomes very heavy. Truck and bus weights are restricted by legislation and it has been found that heavy pleasure cars are exceptionally hard to handle at high speeds.

Within the last year considerable work has been done on what is commonly known as the "tear-drop" car, or a car built with minimum wind resistance by means of stream-lining. By carrying stream-lining to the extreme, real advantages in maximum speed with minimum power can be obtained. While this method of obtaining better performance at high speed is quite promising, the result obtained at low speed is not so satisfactory, and the appearance of a car properly stream-lined to obtain the best results is far from what is commonly accepted as conventional. Any distinct change in the appearance of a car requires considerable pioneering, and there seems to be no way of deciding whether or not this change will be acceptable until after the car is in production for some time.

The object of this article is to point out a way to reduce the weight of motor vehicles by the use of aluminum and in this way improve the performance.

The characteristics of a conventional motor car which affect performance most are outlined by W. S. James.[1] While the curves shown in Mr. James's article and reproduced in Fig. 1, are not based on actual car performance, they indicate in a general way what can be accomplished by weight reduction. These curves are plotted in terms of percent change of the different factors, making the comparative value of each factor very clear.

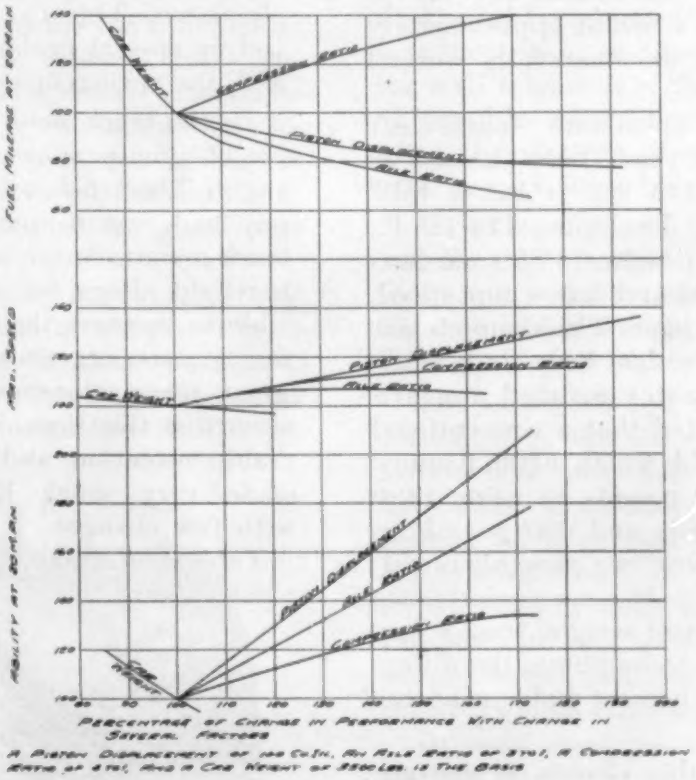


Fig. 1

Weight reduction and an increase in the compression ratio are the only two factors which bring about a positive improvement in car performance without a sacrifice in fuel mileage. An increase in piston displacement or axle ratio improves the performance, but the fuel mileage is reduced. This is especially true for increase in axle ratio which has a very unfavorable effect on fuel mileage and in a much greater degree than an increase in piston displacement. In truck and bus service any decrease in the miles per gallon of fuel would, no doubt, be looked upon with great disfavor.

Most of the vehicles being built today have been increased in size and weight every year since their first introduction to the trade, making it necessary to either increase the size of the motor, increase the compression ratio, piston displacement, or the rear axle ratio to obtain the desired performance. Very little consideration has been given the car weight, yet this factor is very important as can be seen by studying the chart. A decrease of 10% in car weight will result in an increase of approximately 12% in car performance and a reduction in fuel consumption of approximately 20%. Both these improvements would be very acceptable to an owner.

There are very few cars being built today that are not from 20 to 30% too heavy. In most cases this condition can be traced to the lack of appreciation of the advantages of light weight on the part of the general public. If all vehicle owners were as well informed regarding the cost of transportation per pound mile, whether it be for business or pleasure, as commercial vehicle operators are, the demand for lighter vehicles would increase very rapidly.

The most satisfactory way of reducing the weight of a vehicle is by the use of aluminum in both the chassis and body wherever practical. The use of aluminum as an automobile material is about as old as the automobile itself and has been used successfully in most of the im-

*Condensed from a paper presented before the Metropolitan section of the Society of Automotive Engineers, Nov. 1931.

†Chief Engineer, Castings Division, Aluminum Co. of America, Cleveland.

[1] W. S. James. Engine and Car Performance. *S. A. E. Transactions* No. 23.

portant units. It is possible to obtain cast aluminum with physical properties that compare with the ordinary grade of cast iron, or strong heat treated alloys can be obtained which have much higher physical properties. These strong alloys are often used for large crankcases in the more powerful motors, also axle housings, transmission cases, and brake parts. It is also possible to obtain forgings of aluminum which have the same physical properties as mild steel and can be used for front axles, steering arms, connecting rods, or to replace any mild steel forging or steel casting. Sheet aluminum, also rolled structural shapes, for bodies and stampings or tire rims are available with physical properties covering a wide range and adaptable to a very wide variation of conditions.

The cost of aluminum is slightly higher than the cost of more commonly used materials, but by careful engineering this cost can be held to a point ranging from 15 to 25 cents per pound weight saved, depending on the kind of material required, also the design of the parts in question.

The weight saving possible is in most cases approximately 50% of the weight of the iron or steel part replaced by aluminum. In this way a chassis having 1000 lbs. of iron castings in it could be reduced 500 lbs. in weight by the use of aluminum at a cost of approximately $75.00 to $125.00, if the design were carefully worked out.

Fig. 2 shows the chassis of an up-to-date pleasure car which was built within the last year. This chassis, as shown, weighs approximately 2,735 lbs. It has a 146″ wheel base, a 16-cylinder motor, developing 175 H. P., and aluminum is used wherever practical. This car was built for the high-priced car field and has a top speed, fully equipped, of 100 miles per hour. The complete car equipped with a five-passenger sedan body and ready for the road with gas, oil, and water weighed approximately 4300 pounds. It is estimated that a conventional car built of iron and steel would weigh 6500 pounds. This represents a saving of 2200 pounds by using 1420 pounds of aluminum in the chassis and 600 pounds in the body. It is true that one or even two passengers riding in a car of this kind will change the car performance very little, but by removing a ton of weight from a car, as is possible in this car by using aluminum, the difference would be appreciated by the poorest judge of motor car performance.

The general construction of this chassis is conventional in every respect. The motor is an overhead valve "V" type with the crankcase and cylinder blocks in one piece. Removable Ni-Resist iron sleeves were used for the cylinder barrels. The coefficient of expansion of this material being very close to that of aluminum, made it

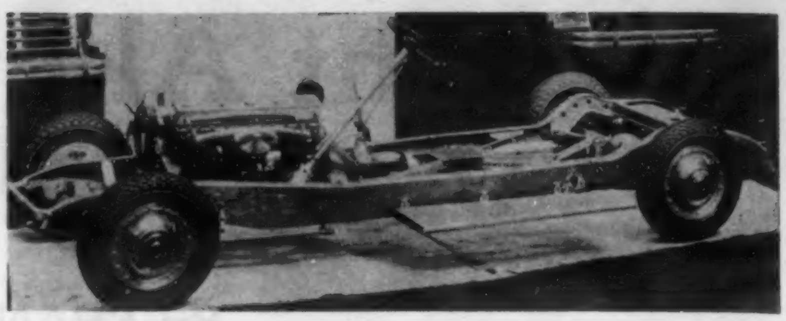


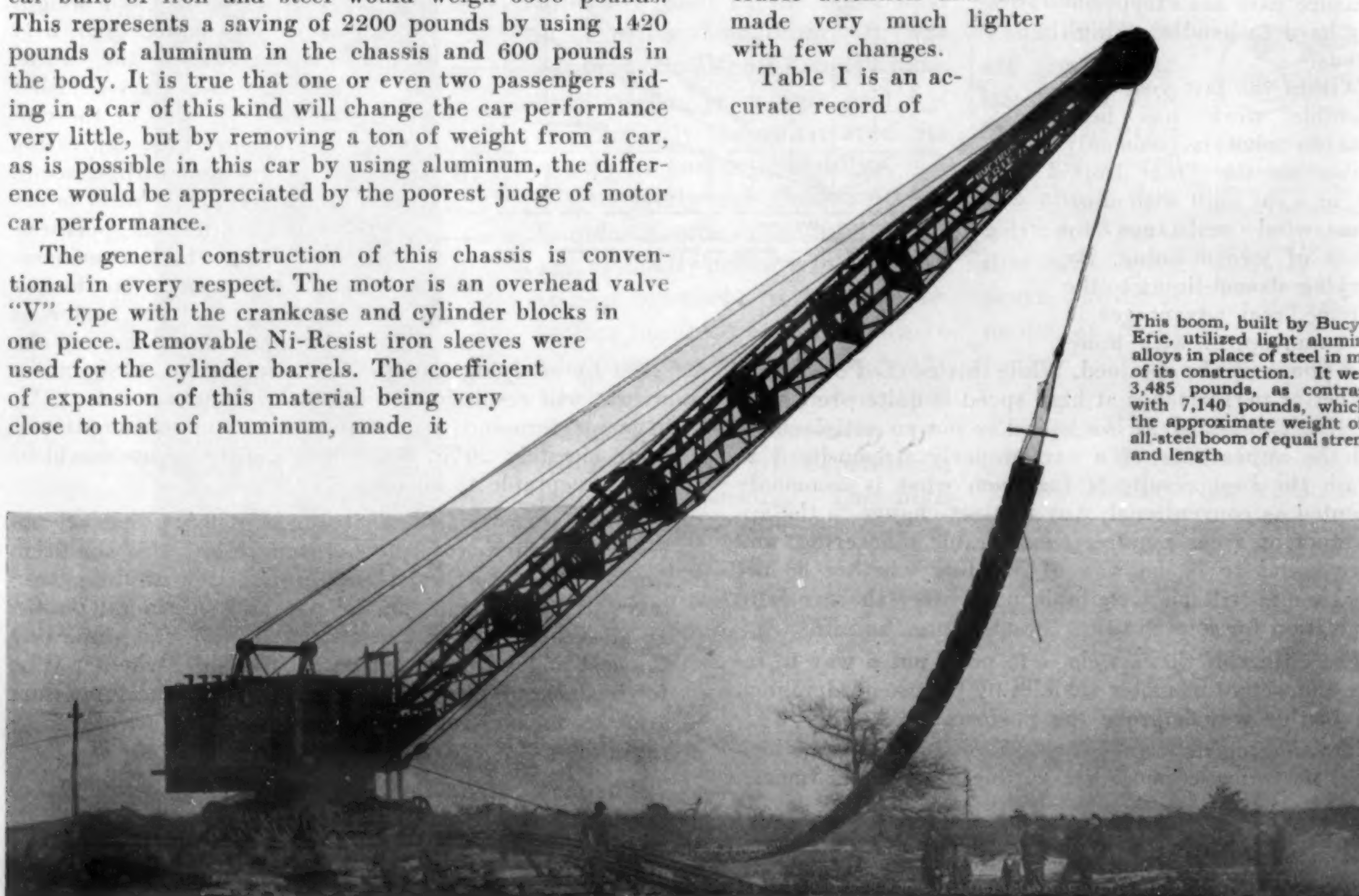
Fig. 2

possible to fit the aluminum pistons very close without any danger of high expansion pressure on the cylinder wall causing excessive wear or seizing. The connecting rods are of aluminum, also the cylinder head with aluminum-bronze valve seats and Ni-Resist iron valve stem guides shrunk in place.

The front axle is an aluminum forging and the rear axle, worm drive type with a cast aluminum housing, differential carrier, brake anchor plates, and brake shoes. The wheels are aluminum castings. The frame is of pressed sheet aluminum and weighs approximately 230 pounds complete. The body is built almost entirely of aluminum. This car handled very well at high speeds and no special problems were encountered in connection with the spring suspension.

In the truck field the advantages to be gained by the use of aluminum are considered from a slightly different angle. The truck operator is most vitally interested in pay load, cost of operation, and cost of maintenance. The truck manufacturer seems to be rather slow about adopting light alloys for chassis construction, making it possible to increase the pay load, but the truck operator is much more aggressive in this respect and there are a great many aluminum tanks and aluminum bodies in service at this time. The trailer is also receiving considerable attention and can be made very much lighter with few changes.

Table I is an accurate record of



This boom, built by Bucyrus-Erie, utilized light aluminum alloys in place of steel in much of its construction. It weighs 3,485 pounds, as contrasted with 7,140 pounds, which is the approximate weight of an all-steel boom of equal strength and length

Table I

Maximum total weight permitted by Pennsylvania "W" License		22,000 lbs.
Weight of chassis, Mack AB 2½-ton	7,260 lbs.	
Weight of steel body, mechanical hoist and high-lift mechanism	4,600 lbs.	
Total Weight of steel body		11,860 lbs.
Weight of aluminum body, mechanical hoist and high-lift mechanism	2,800 lbs.	
Weight of aluminum body	10,060 lbs.	
Payload with aluminum body	11,940 lbs.	
Payload with steel body	10,140 lbs.	
Gain per load		1,800 lbs.
Average number trips per day		10
Gained per day per truck		18,000 lbs.
Number days to gain free one day's work of steel job equals (101,400/18,000) or		5.63 days
Days work gained per year (300 days)		53.3
Operating cost per truck per day		$25.00
Cost of 53.3 days' work saved per truck per year		$1332.50
Added investment for aluminum body and hoist		$ 645.00
Gain in capacity for $645 worth of additional investment		17.7%
Return on additional investment		206.0%

the advantages gained by the use of aluminum in a 6½ cu. yd. capacity, aluminum dump body as compared with the 5 cu. yd. capacity steel body which it replaced. These figures were copied from an article read by F. D. Goll before the Pittsburgh Section of the S.A.E. and indicate a definite return on the investment for additional equipment of 206% which is very attractive to any truck owner.

Figs. 3 and 4 show tank equipment for the transportation of milk. The capacity of these tanks is 1000 and 1300 gallons and they replace glass tanks of smaller capacity, but the same weight. It is estimated that the increased capacity of these tanks insure a saving of $1.85 a day, making it possible to write off the additional cost of the aluminum tank over that of the glass-lined tank in 270 operating days.

In most cases the best way to take advantage of the weight reduction possible by the use of aluminum is to increase the pay load and maintain the same gross load. If, however, the chassis is overweight with reference to the highway limits, aluminum construction will often permit hauling the same load without exceeding these limits.

Interest in aluminum frames is becoming more common and makes it possible to save considerable weight in this one part. The use of aluminum in the rear axle has proved very satisfactory when it is properly designed for aluminum. Both of these parts carry heavy loads and the rigidity of the sections should be given careful consideration to insure satisfactory results. In some cases it is difficult to obtain proper frame depth for satisfactory stiffness, but in most cases there is no real trouble. The rear axle often has to be considerably increased in diameter or reenforced with either an aluminum or steel tube, depending on the available space. When considering changing the material of any iron or steel part to aluminum, it is important that the physical characteristics of both materials be considered. Aluminum is not as stiff as iron or steel when considered section for section. In order to obtain the proper stiffness it is often necessary to add ribs, make the ribs deeper or make the walls of the aluminum part thicker. If the entire casting is made heavier, the weight saving will not be as great as if the added metal is distributed over the points where it will be most effective. Careful design is always productive of better results and cheaper castings. It is also very important to eliminate all sharp corners and make the wall thickness of the parts being changed as uniform as possible. Sharp corners or thin fin-like ribs are very often the starting place for cracks which ruin the part after a very short life.

A combination of the light weight body equipment and the light weight chassis makes pay loads possible that are a real economy and insure a reduction in operating cost which will more than pay the extra cost of this light equipment in a very short time.

The motor bus operator is interested in both the performance of the bus and the pay load. He must have satisfactory speed and acceleration, also a maximum pay load. The short-haul operator usually makes a great number of stops and starts while the long distance operator must have a bus capable of passing slower vehicles on the road to make satisfactory speed, also maintain a fair average speed. Street car operators find that light-weight street cars are a big help in speeding up their schedules and reducing the cost of operation.

In most cases the truck chassis manufacturer is also the bus chassis manufacturer and it is hardly to be expected there would be any great difference in the attitude of these manufacturers toward the light-weight chassis. It is a fact, however, that the light-weight aluminum bus body has been in use for some time and light-weight chassis using aluminum are increasing in number. Conventional iron and steel rear axles, frames,

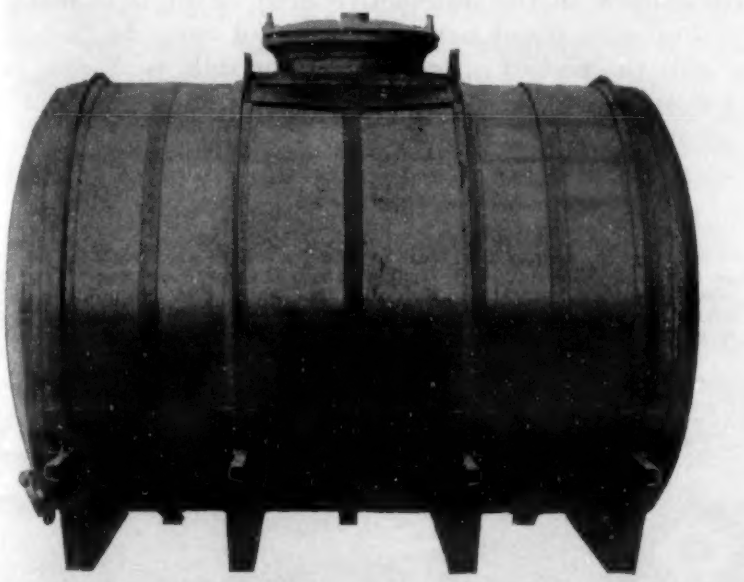


Fig. 3
One of the aluminum tanks for the Fairfield Western Maryland Dairy, showing construction, before applying insulation and lugging.

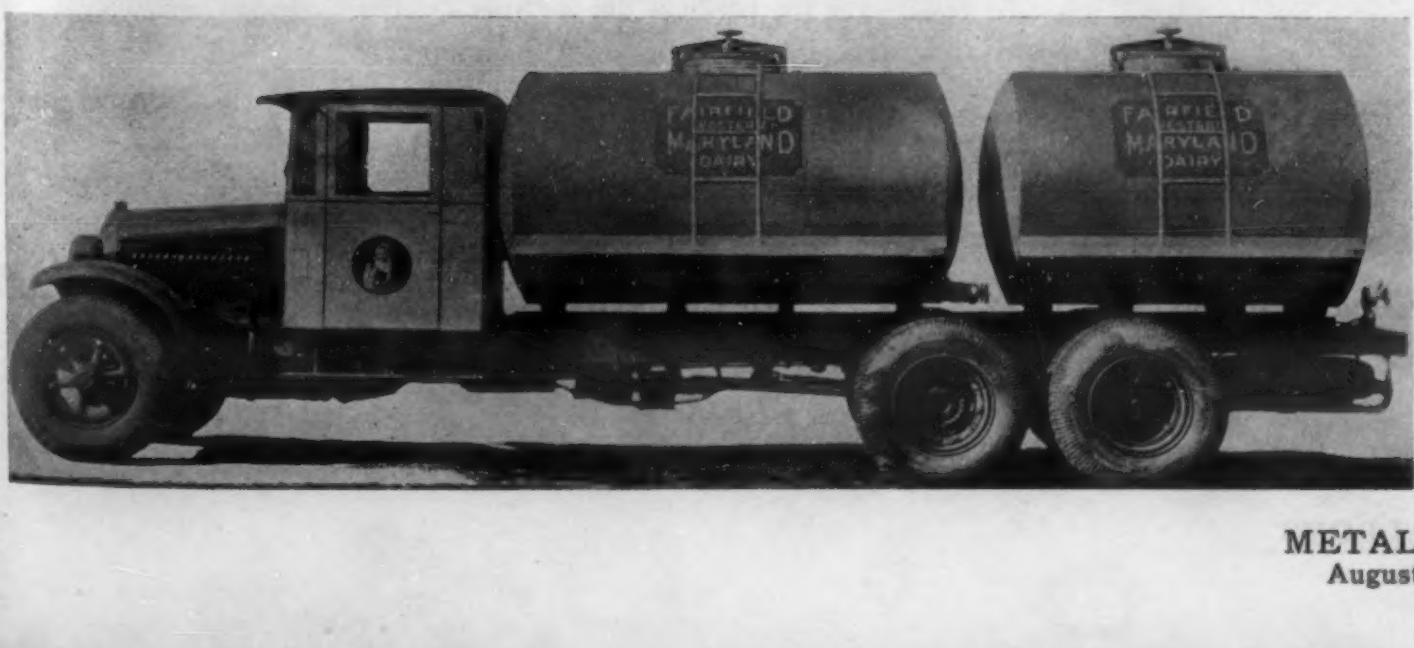


Fig. 4
Two - tank milk transporting unit operated by Fairfield Western Maryland Dairy. Forward tank has a capacity of approx. 1250 gals.; rear tank, 1000 gals. Except for the insulation, both are constructed from aluminum alloys throughout, by the use of which a combined increase in load of nearly 200 gallons was brought about without any increase in gross load.

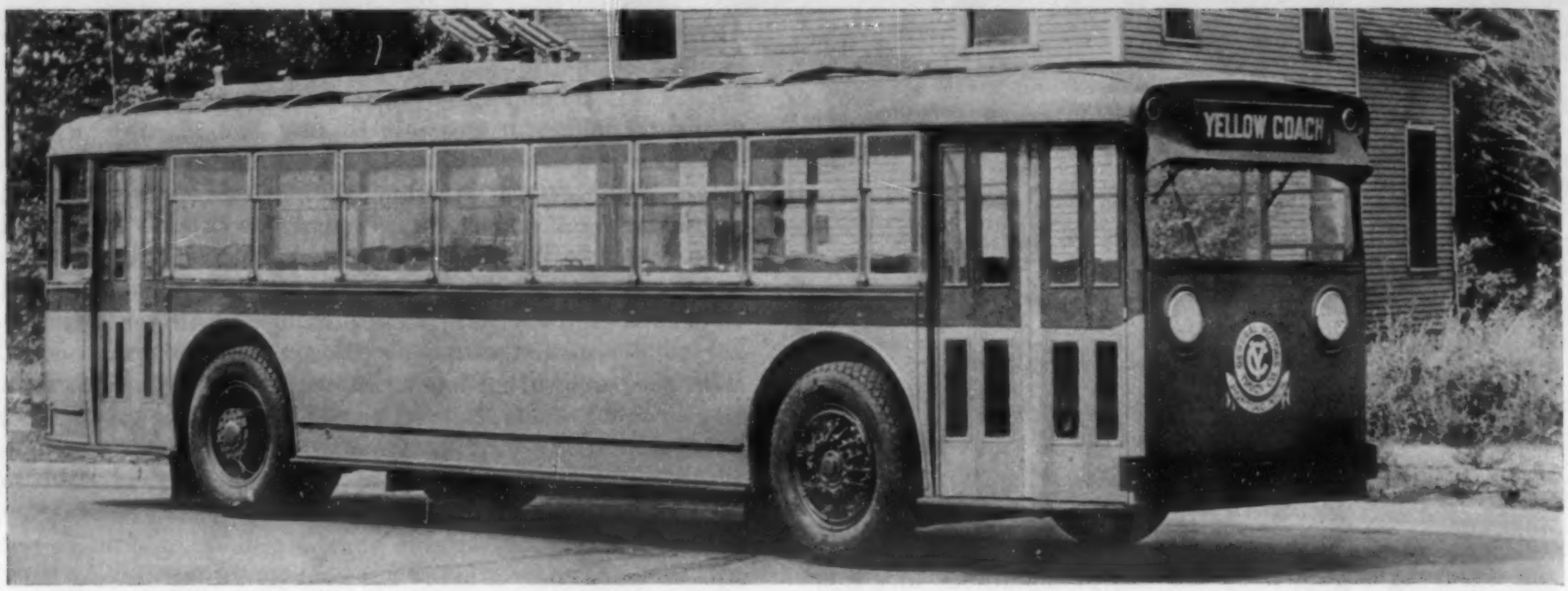


Fig. 5. Light, strong aluminum alloys for body members, frame sections and wheel sections in this Yellow Coach, made possible a saving in weight of two tons, without sacrificing strength or stability. The trolley poles are aluminum and are approximately half the weight of ordinary poles. Their use minimizes trolley jumping and reduces wire wear.

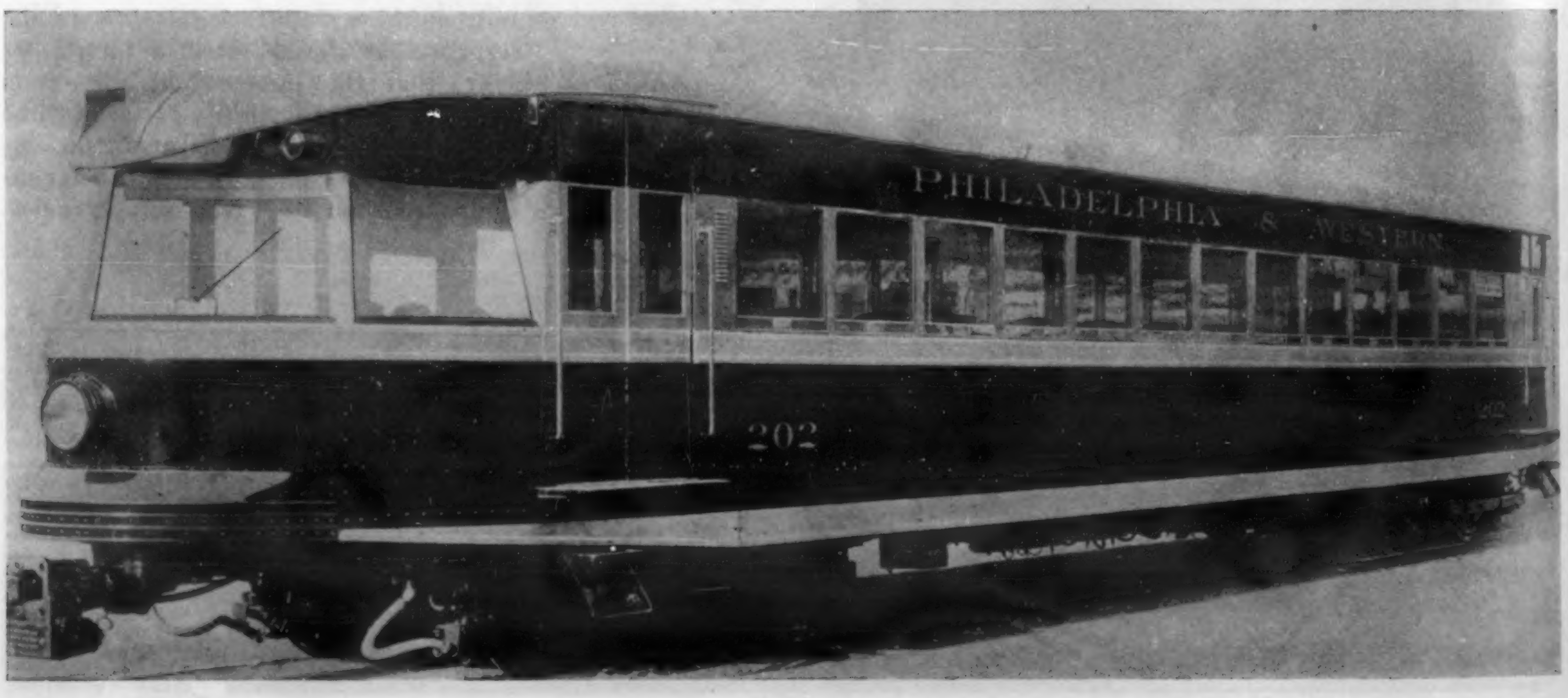
and crankcases are very large in buses, making it possible to save a great deal of weight by adapting aluminum for these parts.

There is a tendency to eliminate the heavy steel frame in the latest bus designs and use a combined body sill and frame side rail to do the work these parts shared in the older designs. The front axles can be made of aluminum forgings which are just as stiff as the steel forgings. When the aluminum axle is designed stiff enough, it will be found to be much stronger than the steel axle in tension due to the fact that the area of the aluminum axle cross section is so much greater than the steel axle cross section and the tensile strength of the aluminum is almost that of the steel. The rear axle should be made of either one piece or built up, depending on the size. The transmission case should be aluminum, also the cylinder block, head, and crankcase. By combining a chassis of this type with a light-weight body, the problem of obtaining pleasure car performance in a bus would be very much simplified.

Fig. 5 shows one of the newest type buses equipped with an aluminum body having the body sill and chassis frame combined. Aluminum has been used wherever practical in this bus, making it possible to reduce the weight of the completed vehicle approximately 4000 lbs. as compared to a conventional vehicle of equal capacity. By the use of 3946 lbs. of aluminum it is possible to eliminate weight equivalent to 26 pay passengers which is a large item when considered in terms of cost of operation and maintenance.

Light weight vehicles are being developed in every line of transportation and while progress in the pleasure car field is rather slow, it is largely due to the fact that most pleasure car owners have very little information regarding the cost of operating their cars. In any section of the transportation field where the cost of operation is accurately recorded, development work in light weight units is progressing very rapidly. Light weight is not being developed in the automotive field only; it is being developed in street cars and railroad cars. In both these fields the cost of operation per ton mile is of vital interest and brings the advantages of light weight out very clearly on a cash basis.

One of the aluminum cars built by the J. G. Brill Company for operation on the lines of the Philadelphia & Western Railway Company. The entire body, including under frame, roof frame, sides, and interior and exterior sheathing is made from aluminum alloys The belt rail and skirt are polished aluminum. The car weighs complete, only 50,000 pounds.



Steel in the Light of the Precipitation Theory

BY ALBERT SAUVEUR

The Hardening of Steel

LET US consider a steel of eutectoid composition:—On entering the quenching bath the steel consists of a solid solution of carbon in γ-iron. While in the quenching bath, on reaching a temperature of some 300° C., the γ-α allotropic transformation occurs very quickly. The resulting α-iron, for a very short period of time at least, retains 0.85% carbon in solid solution and is therefore *excessively* supersaturated. This leads to immediate and abundant carbide precipitation in submicroscopic particles, in other words to aging with its resultant increase of hardness.

Assuming that this aging in the quenching bath has been complete, the α-iron should be quite, if not altogether, free from carbon in solution. Further aging (increased hardness) should not take place at room temperature nor on reheating slightly above room temperature. If aging takes place, it may be due to the fact that the aging in the quenching bath was not complete, or more likely to the presence of undecomposed (retained) austenite.

In case of complete aging in the quenching bath, reheating above room temperature (tempering) should cause agglomeration of the precipitation particles, hence softening, which is in accordance with facts.

Agglomeration reaches its maximum condition in the sorbitic state which results from reheating the quenched steel to some 600° to 700° C. This also corresponds to the maximum softness that can be imparted to quenched steel without reheating it above the critical range when a γ-solid solution is again formed, unless, indeed, the steel be maintained a considerable length of time in the sorbitic range when it will spheroidize, which is an exaggerated case of agglomeration, and become still softer.

This theory leads us to consider martensite as an aggregate resulting from complete aging of a solid solution of carbon in α-iron, its ultimate composition being α-iron and carbide particles in submicroscopic dimensions. The agglomeration of the particles induced by reheating (tempering) produces the troostitic constituent and with larger size particles the sorbitic constituent.

The well known fact that troostite is colored more readily and intensively by some etching reagents than martensite and sorbite must be explained. It might be due to the fact that carbide particles of submicroscopic size (as present in martensite) do not respond readily to the etching attack and that coarsely agglomerated particles (as present in sorbite) fail likewise to be readily acted upon whereas finely divided particles (as present in troostite) are readily affected. The speculative character of this explanation is admitted.

The Annealing of Steel

Let us again consider a steel of eutectoid composition. On cooling it slowly to and through its critical range, the γ-α allotropic transformation occurs at some 700° C. The resulting α-iron, for a very short period of time at least, retains 0.85% carbon in solid solution. Submicroscopic precipitation of carbides, however, follows immediately and the austenite is converted into martensite. At the high temperature prevailing, coagulation sets in and martensite is converted into troostite. This in turn is followed by the formation of larger carbide particles resulting in the production of sorbite. Finally sorbite transforms into pearlite. Following this line of reasoning, the constituents of steel could be described as follows:

Austenite: A solid solution of carbon in γ-iron.

Martensite: an aggregate of α-iron and of carbide particles of submicroscopic size.

Troostite: an aggregate of α-iron and of carbide particles of microscopic size.

Sorbite: an aggregate of α-iron and of larger carbide particles.

Pearlite: an aggregate of α-iron and of carbide exhibiting a eutectic (eutectoid) pattern.

Such a theory is opposed to the views of those who claimed that on slow cooling from above the critical range, austenite decomposes directly into pearlite. The evidences offered in support of that contention, however, are far from conclusive. It is difficult to conceive that the allotropic transformation of the solvent and the precipitation of the solute, the two steps involved in the decomposition of austenite, take place so suddenly and completely that no transition stages exist. Physical transformations are of necessity gradual, even when extremely rapid. Stages must exist when the alloy contains both γ and α iron, and both carbon in solution and precipitated carbide before the transformation is complete. When water freezes mixtures of ice and water must exist before the water is completely frozen. These correspond to our transition constituents.

The author realizes that this use of the precipitation theory to explain the structure of steel is not in accord with some of the views he has expressed in the past—but this is of little moment. The author is not wedded to any theory, being at all times prepared to discard any he may have upheld in favor of a more plausible one. Moreover, this plea for the precipitation theory does not signify that he is accepting it to the extent of abandoning all other explanations of the structure of steel.

The simplicity of the precipitation theory has a strong appeal and the possibility it affords to explain both the structure of annealed and of hardened steel deserves our careful consideration. The author has attempted to make for it as strong a case as possible in the hope that it may induce others to offer criticisms, either in support or in opposition.

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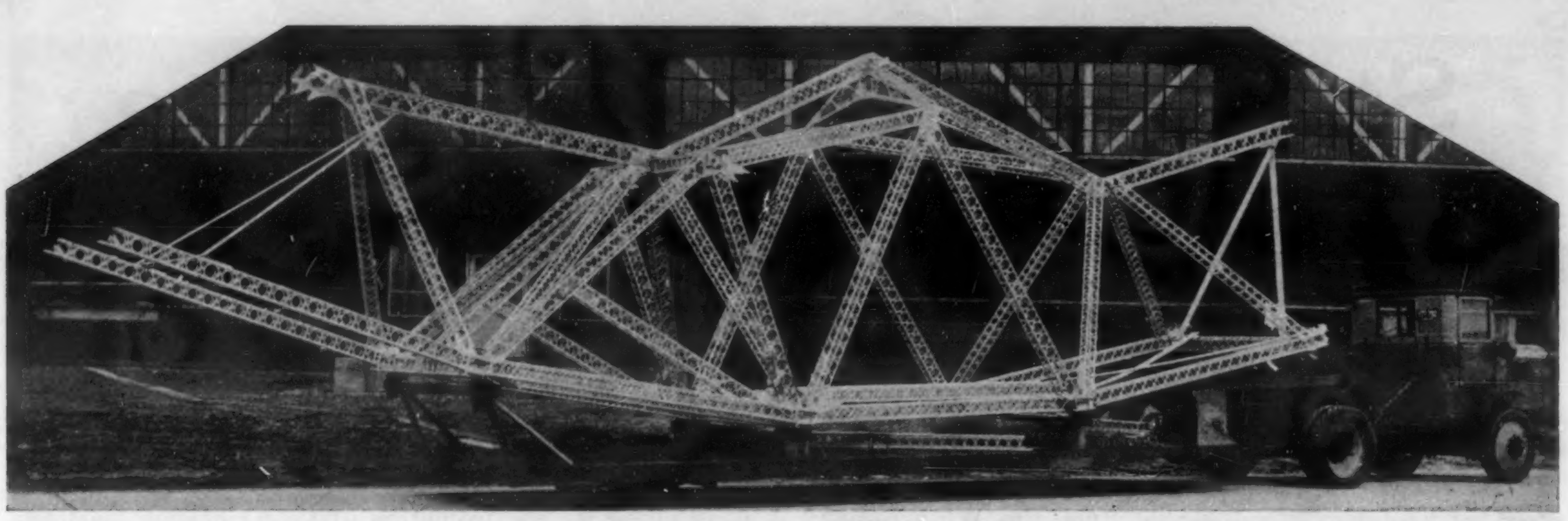
Organization of the **Steel Treating Materials Company** was announced by **Ray Thiesenhusen**, president. Sales offices and warehouse have been established at 1321-1329 West Pierce St., Milwaukee, Wis.

Mr. Thiesenhusen was affiliated with the Wesley Steel Treating Company, as treasurer, sales manager and a member of the board of directors. For the past year Mr. Thiesenhusen was chairman of the Milwaukee Chapter of the American Society for Steel Treating.



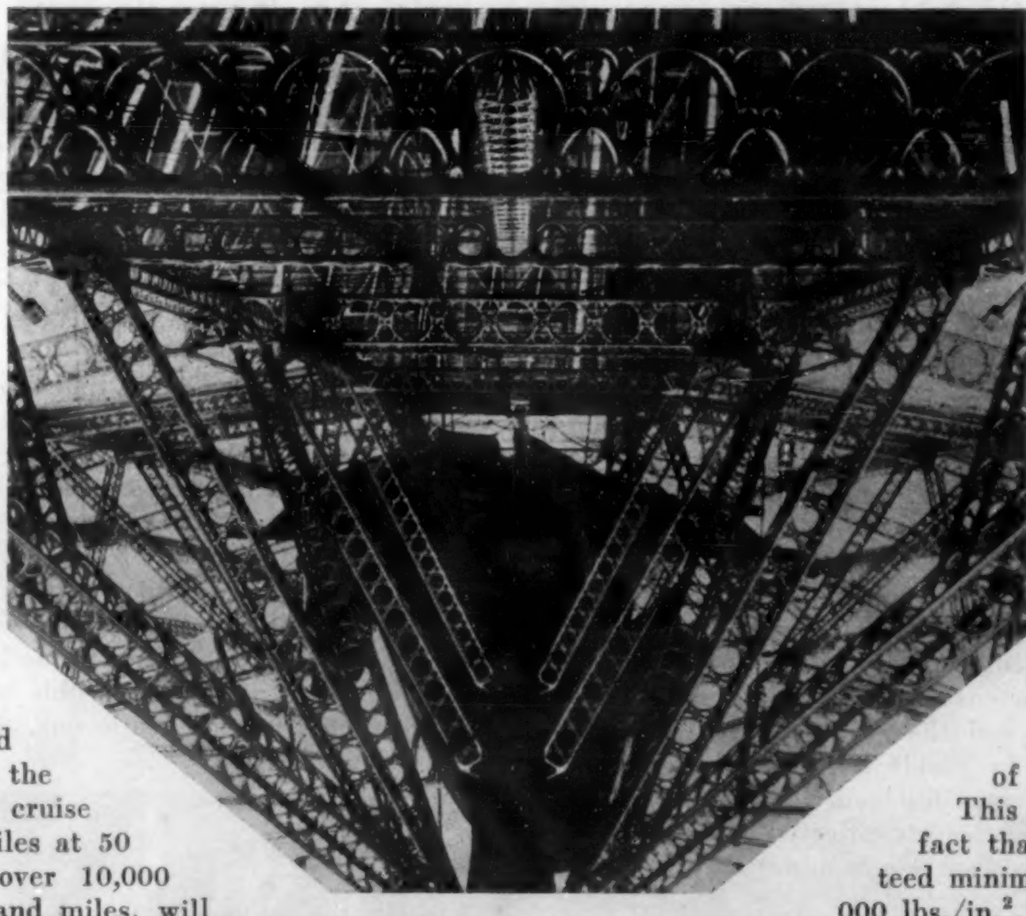
Transporting partially completed frame to dock for final assembly. Girders for main frames and longitudinals were fabricated in a separate plant enabling simplification of assembly operations.

Aluminum Alloys in the U. S. S. AKRON†

By G. O. HOGLUND*

THE SUCCESSFUL performance of the Akron and the present construction of a still larger air ship for the Navy make it worth while to outline some of the characteristics of the structural material used in her, in order that the advances made in this field can be applied to other structures, particularly airplanes.

In order that the magnitude of the job can be appreciated the main dimensions of the Akron will be listed. The nominal gas volume is 6,500,000 cu. ft., as compared to 2,500,000 cu. ft. for the Los Angeles. The length is 785 feet and the diameter 132.9 feet. The estimated dead weight is about 220,000 lbs. The total lift based on the above gas volume is 403,000 lbs. Eight engines, with a total rated horsepower of 4,480 are installed in ventilated engine rooms inside the ship. The Akron will cruise for 9,000 nautical miles at 50 knots. This range, over 10,000 miles expressed in land miles, will enable reaching almost any point on the globe, for, if the distance is too great in one direction the course can be laid around the world in the other direction. It will also permit cruising from the Atlantic Coast to Europe and return, without refueling.



View showing complicated structure along top cat walk. Note flanged lightening holes in girder near top of picture. This forming was done on fully heat treated duralumin.

The structural framework of the Akron is formed almost entirely from aluminum alloy that has been fabricated to obtain a higher yield point than is usually obtained with this material. The type of girder developed by the Goodyear Zeppelin Corporation for the Akron and her sister ship, the ZRS-5 will develop a strength of substantially over 35,000 lbs./in.^{2}, in compression before secondary or form failure occurs. This is a distinct improvement over the strength of the old Zeppelin type of construction as applied on the Los Angeles. The application of ordinary duralumin or 17ST as it is known to the aircraft industry, would not allow full advantage to be taken of the new girder section. This can be seen from the fact that 17ST has a guaranteed minimum yield point of 30,000 lbs./in.^{2}, with an average value of 35,000 lbs./in.^{2}, and is somewhat under the strength which the girder will develop. In addition the most severe conditions on many such girders is tension rather than compression which also makes a high

*Aeronautical Engineer, Aluminum Co. of America.

†Paper presented at a combined meeting of the Pittsburgh Aero Club and the Pittsburgh Section of the Society of Automotive Engineers, Dec. 1, 1931.

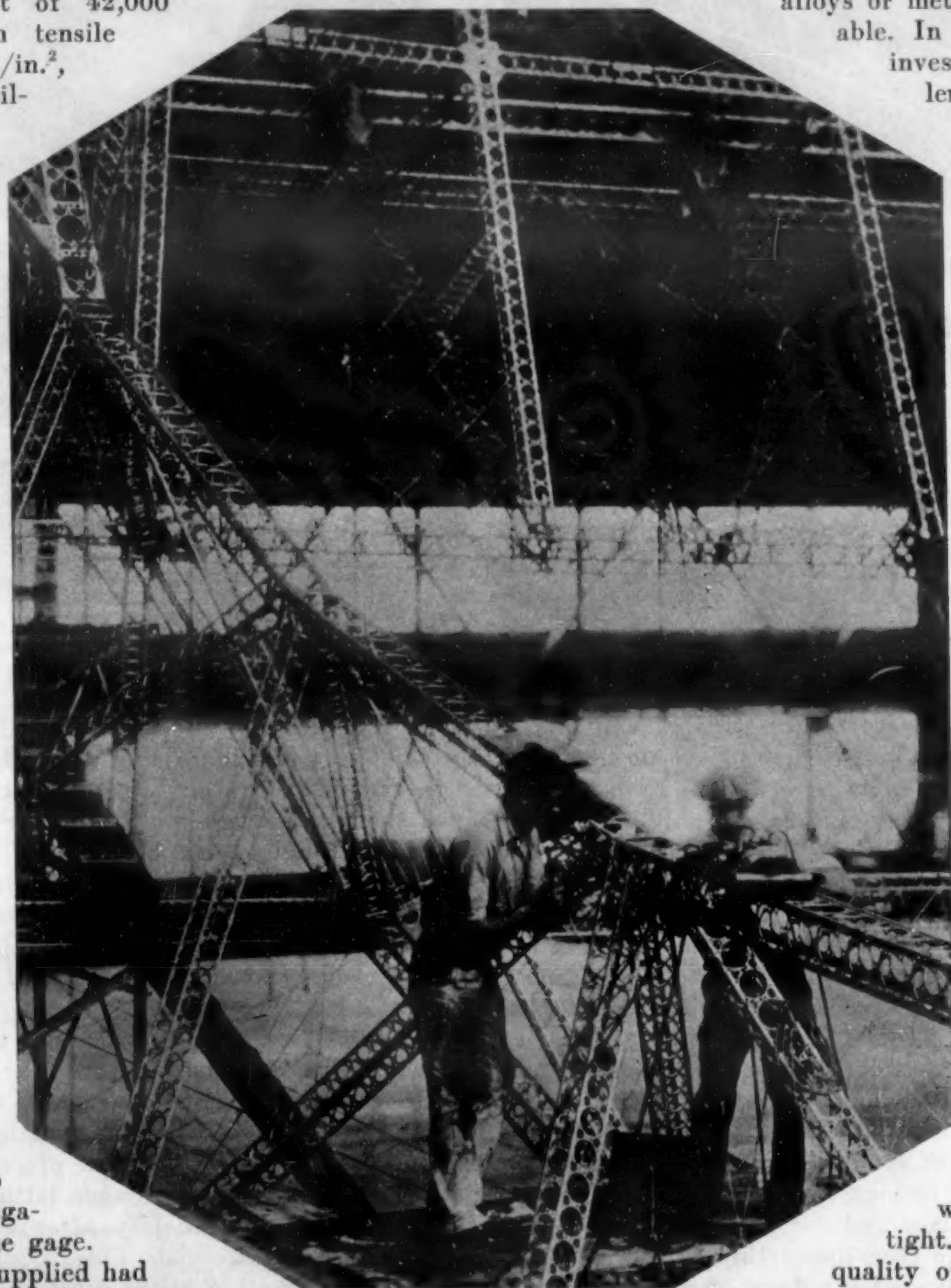
yield point desirable. It may be worth noting here that it is considered that the most rational criterion of the strength of the material in any structure is given by the yield point rather than the tensile strength. This is somewhat at variance with the practices in the airplane field though there have been some recent suggestions to use the yield strength in the design of airplanes.

The problem to be answered then was to provide an aluminum alloy with a minimum yield point of 42,000 lbs./in.^{2}, a minimum tensile strength of 55,000 lbs./in.^{2}, with sufficient workability to permit the necessary forming operations, and with a corrosion resistance equal to or better than 17ST, or ordinary duralumin.

The material actually supplied is known as 17SRT. This is 17ST material which has been given a gage reduction after heat treatment. The composition of this alloy is approximately 4% copper and 0.5% each of manganese and magnesium, with the balance aluminum. The actual physical properties average about 46,000 lbs./in.^{2} yield strength, about 61,000 lbs./in.^{2} tensile strength and from 10% to 15% in 2" in elongation depending on the gage. All of the material supplied had physical properties above the minimum values specified above.

The physical properties obtained on 17SRT were satisfactory to develop the full strength of the girders. The cold rolling after heat treatment affects the workability of the material but this operation was carried only far enough to obtain the combination of mechanical properties desired and still retain a sufficient degree of workability to permit the cold forming operations. The corrosion resistance of the material after cold rolling was thoroughly investigated to determine if this operation had a harmful affect. It was determined that the corrosion resistance of 17SRT is substantially the same as that of 17ST, at least it is not inferior. 17SRT had one other big advantage in that all heat treatment operations were carried out in the sheet mill and this operation was eliminated entirely from the plant operations of the Goodyear Zeppelin Corporation.

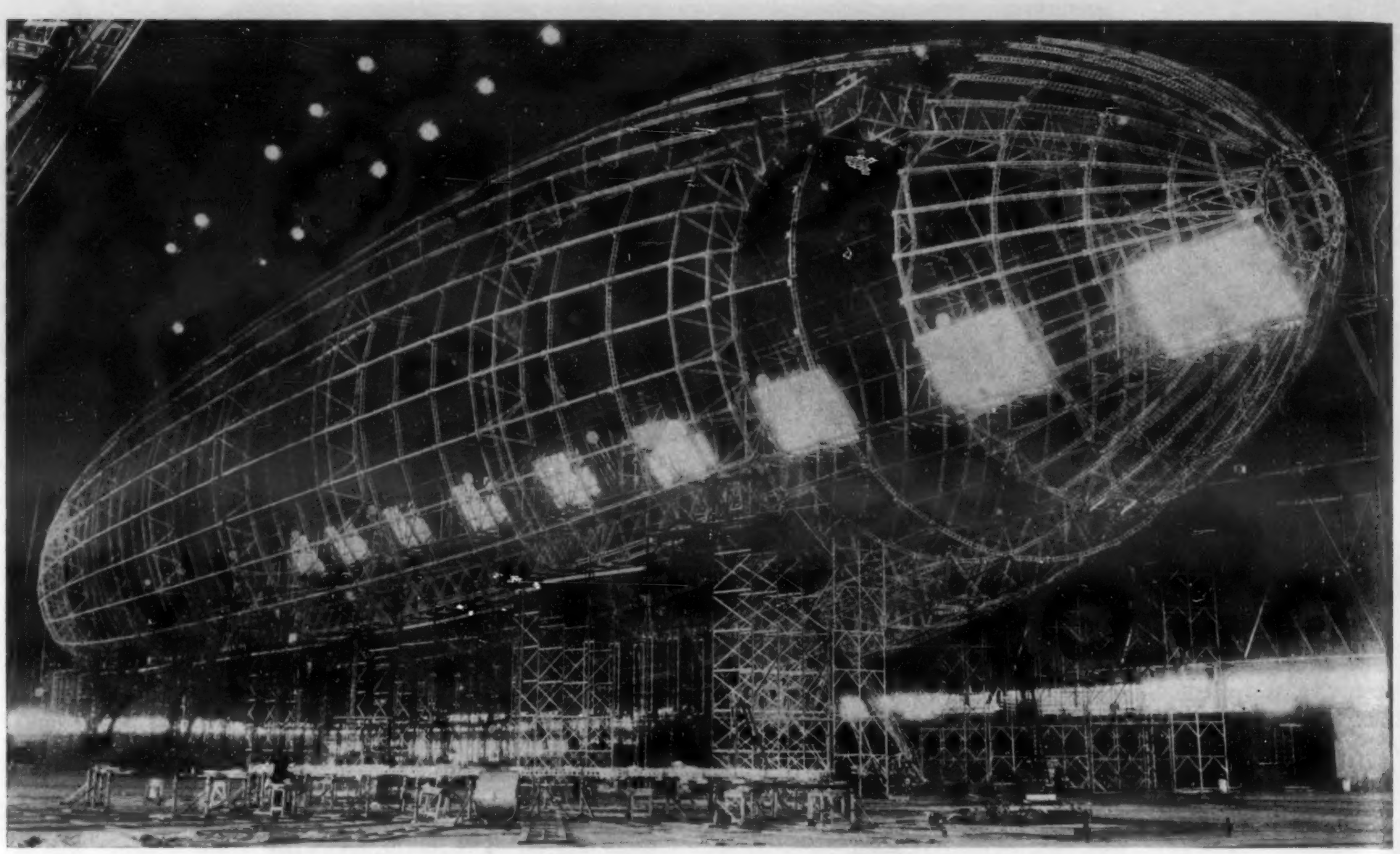
Weight limitations and control on the Akron required exceptionally close gage control of the material. Every piece of material was gaged and held to rigid tolerances on thickness and width. Every lot of material was tested to insure mechanical properties above the minimum and analyzed to determine the constituents.



Completing assembly of main frame and longitudinals of U. S. S. Akron. Note netting for retaining gas cells.

In addition to the 17SRT sheet used in the girders there are parts where other alloys or metals proved more suitable. In many of these cases investigation of the problem indicated the most suitable material, considering strength, lightness and corrosion resistance, to be an aluminum alloy. For example, the engines have aluminum alloy castings for crankcases, valve housings, pistons and other parts. Heat treated aluminum alloy castings were used for such parts as control panel handles, ballast bag release valves, shut off cocks, safety valve parts, rudder and elevator hinge parts, indicator supports and controls, lighting fixtures, valve parts, mooring winch parts and other parts. In most cases sand castings were used and if necessary were made pressure tight. In many cases the quality of these cast parts was investigated with X-ray equipment to insure sound parts. Aluminum alloy forgings also played an important part in the construction of the Akron, being used for such parts as valve bodies, pulley and sheaves, fuel pump parts, etc.

Perhaps more positive steps were taken to prevent deterioration of the structure on the Akron than on any previous airship. The sheet after forming was cleaned, a chromic acid anodic oxide coating applied to prepare the metal for painting and a coat of aluminum paint applied before the parts were assembled. A second coat of paint was applied after the girders had been assembled. Considerable experimental and practical experience has been compiled on painting the aluminum alloys and it can be stated that the above method is one of the best that is known today. An "Extra Fine Lining" grade of aluminum powder was used in mixing the paint for the



Three quarters front view of U. S. S. Akron showing completed nose-piece swung into position and ready for tying in to main structure.

structure while the fabric cover was treated with dope mixed with a "Standard Varnish" grade of powder.

About 6,500,000 duralumin rivets were used in assembling the Akron, as well as a considerable number of duralumin screw machine products such as bolts, nuts, machine screws and spacers. The fact that so many rivets were driven in the structure made it worthwhile to use a special rivet filleted under the head and designed to save some weight over standard button head rivets. Special riveting tools for squeeze type riveting operations were also designed and used. The rivets were heat treated and quenched by the Goodyear Zeppelin Corporation and driven within half an hour after heat treatment to facilitate forming a perfect head.

The fuel and ballast system of the Akron utilizes aluminum tubings and fittings for fuel lines connecting the tanks and the engines and for dumping and transferring water ballast to trim the airship in the air. All of this tubing was inspected for flaws, dimensions, and in addition, pressure tested in the mill for leaks, before shipment to Akron. The gasoline storage system is made of 120, 365 and 400 gallon aluminum tanks distributed along the two side corridors to keep the ship properly balanced. These tanks are cylindrical in shape and are made from 2S material welded at the joints. Each of the eight engines on board is housed in an engine room inside the airship and lubricating oil is supplied from 200 gallon aluminum welded oil tanks located near each room. The normal gasoline supply of the Akron is about 124,000 lbs. and is distributed in 110 tanks. The piping system permits fuel to be taken aboard at the bow or near amidships for distribution to any of the tanks.

It is interesting to compare the Akron with the only other rigid airship of the Zeppelin type that has been built in the United States, the Shenandoah. In the first place, the material used in the Shenandoah was 17ST, which does not have as high mechanical properties as 17SRT used for the Akron. The higher properties contribute towards increasing both the performance and the safety of the new airship. In addition, the girder section of the Shenandoah was triangular in shape, made up of channel longitudinals braced with diagonal lattice members. As has been explained, the girder section developed by Goodyear Zeppelin Corporation is an improvement over the above in strength and represents a distinct advance when considering mass production and shop practices.

The contour and dimensions of the short lattice members on the Shenandoah were fabricated by forming the lattices before heat treatment and quenching. Distortion in these sections was minimized by heat treating small bundles of the lattices but the process was slow and relatively expensive. The longer lattices and longitudinals were formed immediately after heat treatment before the material had aged. This required coördination between heat treating of the sheet and forming the section, which of course is not the most convenient arrangement. Research programs and practical experience have since determined that the corrosion resistance of fully heat treated and aged 17ST which has been cold worked by either forming or rolling is at least equal to material which is not cold worked after aging. This enabled the use of simplified practices on the Akron.

The individual parts for the box girders on the Akron were rolled to the form desired, blanked out, flanged, on fully heat treated metal. Production schedules and shop operation could be arranged without considering heat treatment in the plant. When the magnitude of the job is considered it is evident that the above coupled with the excellent forming technique developed by Goodyear, enabled the completion of the Akron as early as was done after the letting of the contract.

In conclusion I wish to thank the Goodyear Zeppelin Corporation for the pictures which are shown.

The Constitution of the COPPER-TIN-LEAD SYSTEM

A CORRELATED ABSTRACT By B. BLUMENTHAL*

AMONG THE bronze bearing metals used in engineering practice the Cu rich Cu-Sn-Pb alloys seem to have the special ability to give a satisfactory bearing metal under adverse conditions. Cu-Sn-Pb alloys are used where a metal is exposed to friction (and therefore to wear) under high specific pressure, with shocks and vibrations and at elevated temperature. Such conditions occur in

- hot rolling mill bearings (shocks and heat)
- cold rolling mill bearings (high specific pressure)
- locomotive tender and car journal bearings (especially shocks and vibrations)
- bearings of airplane landing wheels (shocks)
- connecting rod bearings (oscillatory load)
- gate valves in pipe lines (heat and insufficient care)

Some time ago French, Rosenberg, Harbaugh and Cross[1] studied the influence of Pb on the mechanical, bearing and wearing properties of Cu-Sn alloys. This important investigation was carried on without an exact knowledge of the constitution of the Cu-Sn-Pb system. Recently some papers have been published in Germany on the constitution of Cu-Sn-Pb alloys, so that it now seems possible to give a critical survey of the constitution of the system.

*Berlin-Dahlem.

[1] French, Rosenberg, Harbaugh & Cross. *Bureau of Standards, Journal of Research*, Vol. 1, 1929, page 343.

[2] Bornemann & Wagenmann. *Ferrum*, Vol. 11, 1913-1914, page 276.

[3] Giolitti & Marantonio. *Gazetta chimica italiana*, Vol. 40, I, 1910, page 51.

[4] S. Briesemeister. *Zeitschrift für Metallkunde*, Vol. 23, 1931, page 225.

[5] W. Guertler & W. Leitgebel. Vom Erz zum metallischen Werkstoff. Leipzig, 1929, page 391.

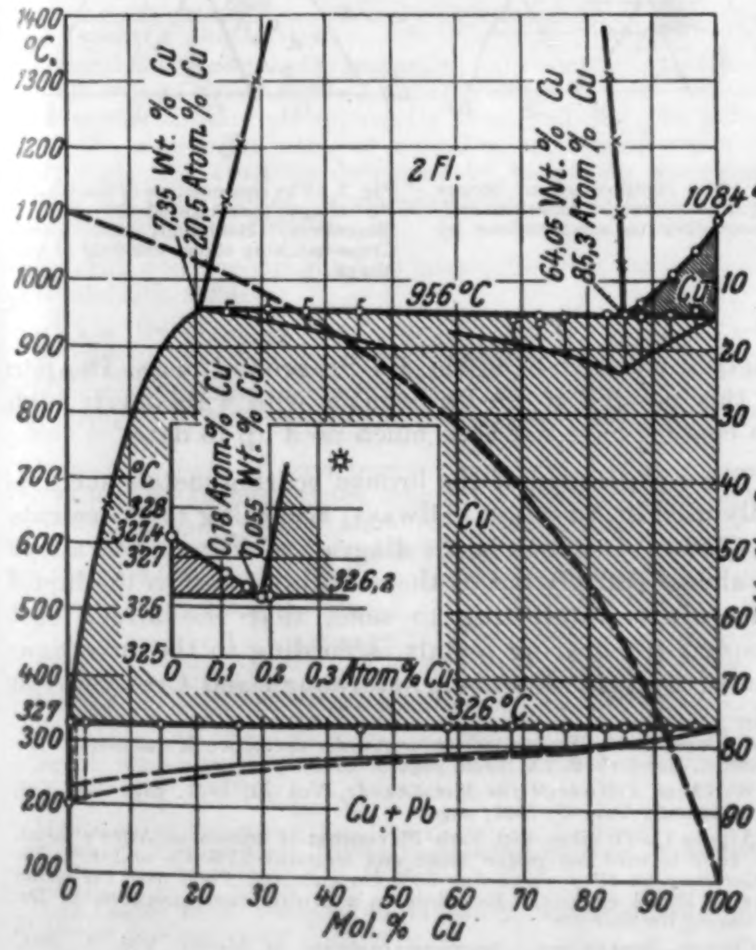


Fig. 1. Phase diagram of the binary Cu-Pb system by Bornemann and Wagenmann.

Fig. 1 shows the phase diagram of the binary Cu-Pb system according to Bornemann and Wagenmann.[2]

The constitution of the Cu-Sn-Pb system was first investigated by Giolitti and Marantonio.[3] They examined only the constitution of the Cu rich alloys. The miscibility gap in the liquid state, which may be derived from the results of their thermal analysis (Fig. 2) is in accordance with the recent examination of Briesemeister[4] (Fig. 3). Guertler[5] found the miscibility gap to be smaller than that of the other investigators (Fig. 4).

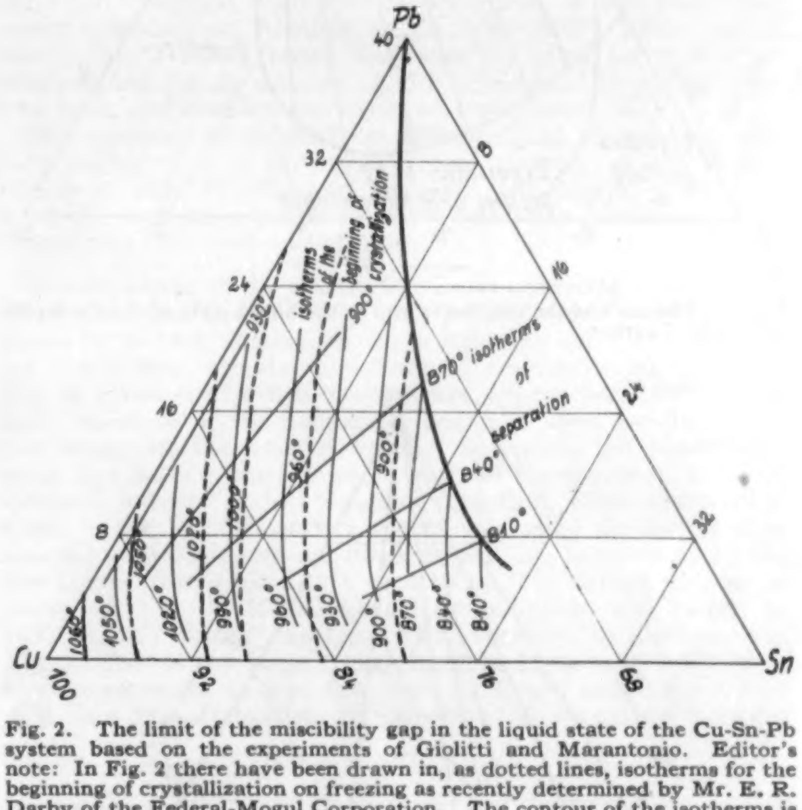


Fig. 2. The limit of the miscibility gap in the liquid state of the Cu-Sn-Pb system based on the experiments of Giolitti and Marantonio. Editor's note: In Fig. 2 there have been drawn in, as dotted lines, isotherms for the beginning of crystallization on freezing as recently determined by Mr. E. R. Darby of the Federal-Mogul Corporation. The contour of the isotherms is very similar, but there are slight discrepancies as to the temperatures. Darby found slightly lower temperatures in tests on 20 lb. melts than in those on 5 lb. melts. Darby states that up to about 5 % Zn, the Zn may be counted as Sn in these alloys in so far as the freezing temperature is concerned.

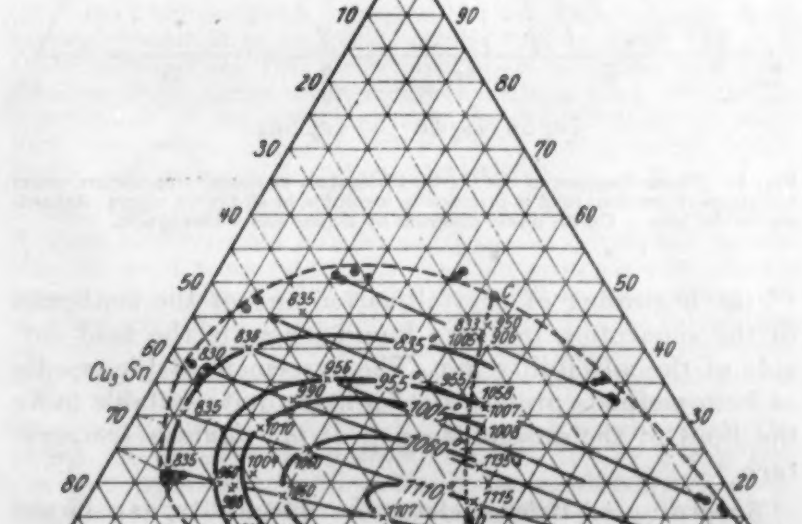


Fig. 3. The miscibility gap of the Cu-Sn-Pb system by Briesemeister.

The methods of investigation used by Briesemeister and by the author based on the experiments of Giolitti and Marantonio were quite different. Briesemeister took samples for chemical analysis out of the molten metal. On the other hand (Fig. 2) the miscibility gap in the liquid state results from extrapolation of the isotherms

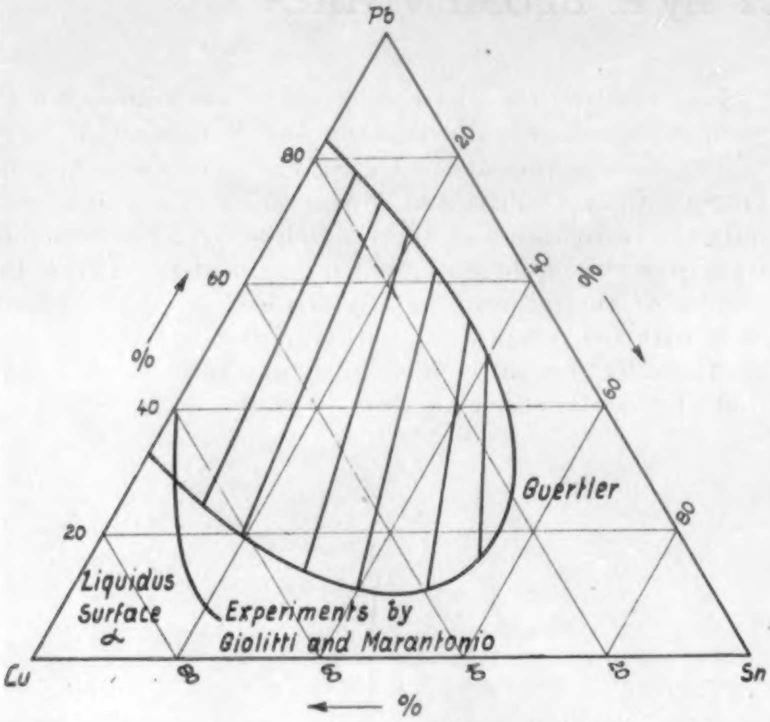


Fig. 4. The limit of the miscibility gap in the liquid state of the Cu-Sn-Pb system by Guertler.

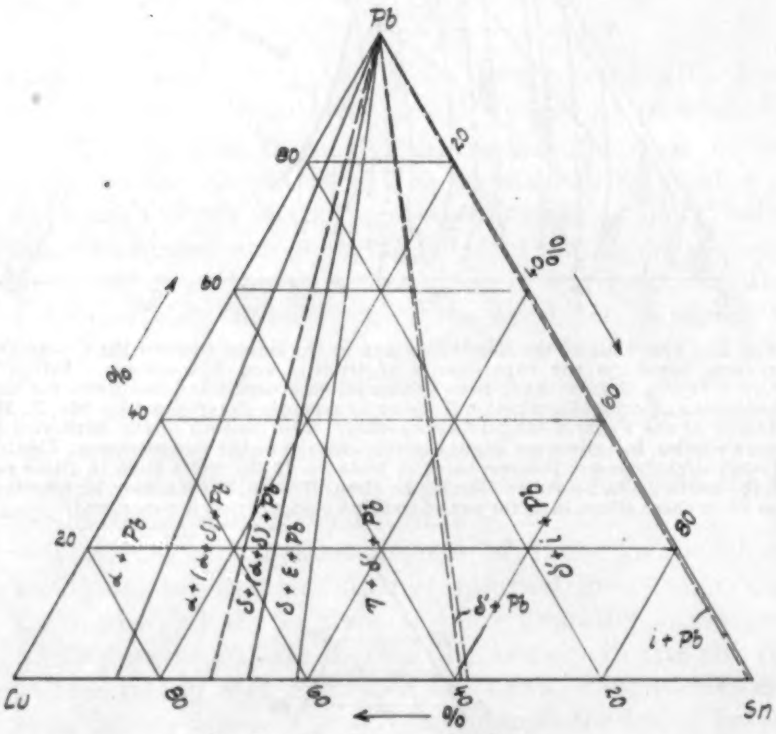


Fig. 5. Phase diagram of the Cu-Sn-Pb system at room temperature under the supposition that lead is practically insoluble in all Cu-Sn alloys. According to the binary Cu-Sn phase diagram by Bauer and Vollenbruck.

of the beginning of crystallization and of the isotherms of the separation into two liquid layers in the field outside of the miscibility gap. The agreement of the results of both methods proves that both are quite suitable to fix the limit of the miscibility gap *at the liquidus temperature.*

According to Briesemeister the miscibility gap in the liquid state is closed at 1130°-1140°C. This is contradicted by Bornemann and Wagenmann who found by experiments based on very careful measurements of electrical conductivity that the miscibility gap in the binary Cu-Pb system is closed at much higher temperature. Similar temperatures should be assumed for the limit of the miscibility gap in the ternary Cu-Sn-Pb system.[6] The cause of this discrepancy is to be found in the fact that the method of investigation used by Briesemeister does not permit of getting exactly the limits of the miscibility gap at temperatures higher than the liquidus temperature. At higher temperatures there may be in the field of the miscibility gap, as is especially emphasized by Claus,[7] a distribution of the Pb in a very finely divided state in the molten metal, so that Pb deposits very slowly. The homogeneous melt really becomes heterogeneous at the high temperatures found for the Cu-Pb system by Bornemann and Wagenmann. The limits found by Briesemeister may suggest those temperatures at which the velocity of deposition reaches a considerable value, but the author does not wish to say that this process is exclusively dependent upon temperature.

The phase diagram of the Cu-Sn-Pb alloys at room temperature is given by Fig. 5.

The alloys lying in the field of the miscibility gap in the liquid state, forming at a sufficiently high temperature a heterogeneous but slowly separating mixture of small drops of Pb in the basic melt, can be used in engineering practice. In pouring the melt it is necessary to take care to pass rapidly through the range of temperature, in which the drops tend to gather and the melt tends to separate into two layers. In the solid state there is obtained a mechanical mixture of drops of Pb in a very

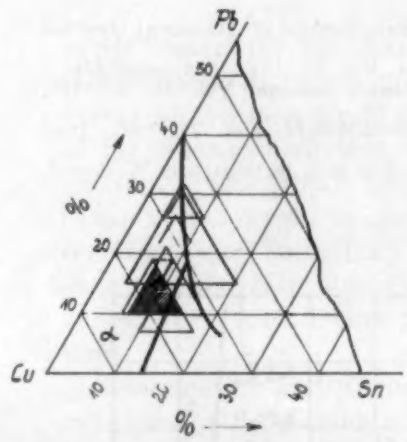


Fig. 6. The composition of bronze bearing metals used in American railways according to specifications by Clamer.

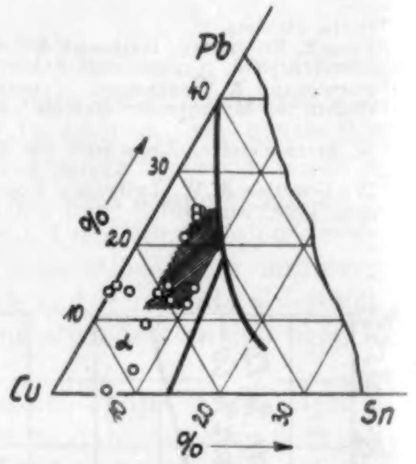


Fig. 7. The composition of the Cu-Sn-Pb alloys investigated by French, Rosenberg, Harbaugh and Cross. Cross-hatching shows the field of best alloys.

finely divided state in the Cu-Sn ground mass. In spite of this possibility Sn-Pb bronzes with a relatively high Pb content have not been much used up to date.[8]

The composition of the bronze bearing metals successfully used in American railways, according to statements by Clamer[9] is shown in the diagram of Fig. 6. The alloys lie almost entirely out of the miscibility gap in the liquid state. It is interesting to note, that the alloys best adapted for bearing metals, according to the investigation of French, Rosenberg, Harbaugh and Cross, fall in

[6] O. Bauer & M. Hansen. *Mitteilungen der deutschen Materialprüfungsanstalten*, Sonderheft IX, 1929, page 6.

[7] W. Claus. *Zeitschrift für Metallkunde*, Vol. 23, 1931, page 264; Kolloidzeitschrift, Vol. 57, 1931, page 14.

[8] A pure Cu-Pb alloy with high Pb content is known as Allen's metal. The alloy is used for piston rings and contains 55% Cu and 45% Pb. In Germany an alloy containing 50% Cu and 50% Pb is used for piston rings of Diesel engines. (According to a private communication of Dr. W. Claus, Berlin.)

[9] Clamer. *Transactions American Institute of Metals*, Vol. 9, 1915, page 241.

(Continued on Page 188)

Structure of Heat Treated Low Carbon Steel

By WELTON J. CROOK* and EDMUND C. BABSON†

DURING THE last few years at Stanford University, work has been carried on relative to various phases of the subject of the heat treatment of low carbon steels. Prior to about 1925, the subject had received but little attention on the part of metallurgists. The lack of interest in this phase of heat treatment was probably due to the fact that most of the earlier attempts to improve the tensile properties of low carbon steels by quenching and drawing gave negative or unpromising results, principally because of the low manganese content and incompletely deoxidized condition of the steels used for experimentation.

The general conditions necessary for the successful heat treatment of low carbon steels, and a record of some of the results which may be obtained, were given in an article by Welton J. Crook.[1] A previous article by Crook and H. S. Taylor[2] dealt with the metallographic changes which take place when a 0.20% carbon steel was drawn at various temperatures after water quenching. H. B. Pulsifer[3] dealt mainly with the practice involved in the heat treatment of bolts and rivets. Pulsifer reviewed the previous literature, including the articles mentioned above.

This article herewith presented has to do particularly with the work of Crook and Taylor, who traced the structures obtained when a 0.20% carbon steel was subjected to a series of water quenches varying from 1250° to 1800°F., and also the structures obtained by progressively drawing specimens of the steel which had been quenched from 1550°F. As a result of this procedure, they advanced the theory that, on heating a low carbon steel, the pearlite changes to martensite at the Ac_1 point and martensite to austenite at the Ac_2. This conception was admitted to be contrary to the theory now accepted, but certain evidence seemed to justify the conclusions reached.

Crook and Taylor also found the following sequence of structural changes to take place when quenched low carbon martensitic steel was subjected to reheating to just above the Ac_1 temperature:

(a) Low carbon martensite.

(b) At a low draw temperature . . ., martensite begins to transform to sorbite.

(c) At 600°F., free spheroidal cementite appears (at 1000 magnification) with the probable formation of a corresponding proportion of free ferrite.

(d) As the draw temperature is increased, more and more spheroidal cementite and free ferrite appear with a corresponding diminution of the sorbite.

(e) At about 1340°F. the steel consists only of free cementite and ferrite.

These deductions may be reconciled with conventional theory. The absence of recognizable nodular troostite is to be expected because this structure is apparently developed only as a quench structure, even when steels of much higher carbon content are dealt with. The troostite developed by tempering martensitic steel is not generally of the nodular form, but has the appearance of "black" martensite.

Jefferies and Archer[4] state, in connection with the tempering of martensitic steel:

> At a temperature near 600°C. (1110°F.) particles of cementite visible under high magnification (500 diameters) appear if the time of heating is sufficient. At the higher temperatures the structure is referred to as sorbite, but there is no sharp dividing line between sorbite and troostite.

In another part of their paper, Crook and Taylor discussed the structural changes which take place when a 0.20% carbon steel is gradually heated from room temperature to 1700°F. Their deductions were based on the structures retained in specimens which were rapidly quenched in cold water.

In this connection they state:

> At a temperature between 1450°F. (787°C.) and 1500°F. (815°C.) an abrupt change in structure occurs, the martensite produced by quenching from 1500°F. (815°C.) having an entirely different appearance from that produced by quenching at 50°F. lower temperature. This point of change apparently coincides approximately with the Ac_2 point, which was determined to be at 1462°F. (793°C.).

> Except for the fact that well established modern theory points to a contrary conclusion, this might be interpreted to mean that when the steel is heated at temperatures above Ac_1 and below Ac_2, the carbon bearing areas are changed to martensite, which is retained on quenching in cold water. On the other hand, if the temperature exceeds the Ac_2 point, the carbon bearing areas are changed through martensite to austenite and a recrystallization takes place in so doing. Water quenching, however, is not sufficiently fast to retain the austenitic structure, so that a martensitic structure of a different form is observed.

The steel used in this investigation was from the bar used by Crook and Taylor, being ¾" round bar stock containing:

Carbon	0.20%
Manganese	0.59%
Phosphorus	0.016%
Sulphur	0.042%
Silicon	0.13%
Copper	0.29%
Nickel	0.19%
Chromium	Trace

This steel was "regenerating," giving a Brinell hardness of 444 when quenched from 1500°F. Specimens of this steel were water quenched in running water from 1300°, 1320°, 1340°, 1360°, 1380°, 1400°, 1420°, and 1450°F., after being held at temperature for 30 minutes. A 5% nitric acid in alcohol etch was used, and examination made at 1000 diameters.

The specimen quenched from 1340°F. (just above Ac_1) was then subjected to a progressive draw starting at 300°F. and ending at 1300°F. with 100° intervals. The time at draw temperature was 40 minutes in each case. The specimens were etched and examined as before.

The structure of the martensitic areas occurring in the specimens quenched from 1340°, 1360°, 1380°, and 1400°F. were found to be very similar, having a light tan color and presenting the typical acicular structure of martensite, as shown in Fig. 3. When the quench temperature approached 1400°F., the light tan color of the martensite began to show localized dark tan areas. In the 1420°F. quench specimen, the martensitic areas had become black in some parts of the specimen, but still retained a more or less acicular structure. Some martensitic areas in the 1420°F. quench specimen showed an abrupt division between dark tan and black coloration; in other areas the two colors were graded into each other. The abrupt division is shown in Fig. 1. If the quench temperature was raised to 1450°F., only "black" martensite was obtained, as illustrated in Fig. 2. This is the same transformation from a light tan to a black martensite as was described by Crook and Taylor, who gave the transformation temperature as occurring between 1450° and 1500°F.

DISCUSSION.

In order to understand the nature of the structures obtained in this steel, it is necessary to investigate the mechanism of martensite formation. Einar Ohman[5] states a belief that martensite is a super-saturated solution of carbon and α-iron. Pairs of carbon atoms replace some of the iron atoms at the lattice corners and deform the body-centered cubic α-iron lattice to a body-centered tetragonal lattice with an axial ratio varying from 1.03 in an 0.08% carbon steel to about 1.07 in a 1.60% carbon steel. This martensitic phase is formed at a temperature which varies with the carbon content, but is practically independent of the quenching rate. In a 0.025% carbon steel this temperature is about 825°F.; in a 0.54% carbon steel, 520°F.; and in a 0.97% carbon steel, 390°F.

Since tetragonal martensite is unstable at all temperatures, it tends to decompose into an aggregate of α-iron and cementite. At room temperature this reaction proceeds very slowly, on account of the rigidity of the steel; but as the temperature is raised, the speed of reaction increases to such an extent that at 210°F. the decomposition is practically complete after two hours. On account of the comparatively high temperatures at which it is formed, martensite has a tendency to decompose on subsequent cooling, even though this cooling is very rapid. The higher the temperature of formation, the greater is this tendency towards decomposition. For instance, when a 0.97% carbon steel is quenched, the martensite is formed at about 390°F. and there is very little opportunity on subsequent cooling for its decomposition; in a 0.20% carbon steel, however, the martensite is formed at about 700°F. and the opportunity for decomposition is much greater. In this case, then, the latter might be expected to have decomposed to a greater extent than the former.

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†Graduate Student, Dept. Mining Engineering, Stanford University.

[1] *Metal Progress*, Vol. 18, Oct. 1930, pages 47-52.
[2] *Metals & Alloys*, Vol. 1, June 1930, pages 539-543.
[3] Low Carbon Steels. *Technical Publication* No. 436, *American Institute of Mining & Metallurgical Engineers*.
[4] The Science of Metals. McGraw-Hill Book Co., 1924, page 381.
[5] X-Ray Investigations on the Crystal Structure of Hardened Steels. *Journal Iron & Steel Institute*, Vol. 123, 1931, pages 445-462.

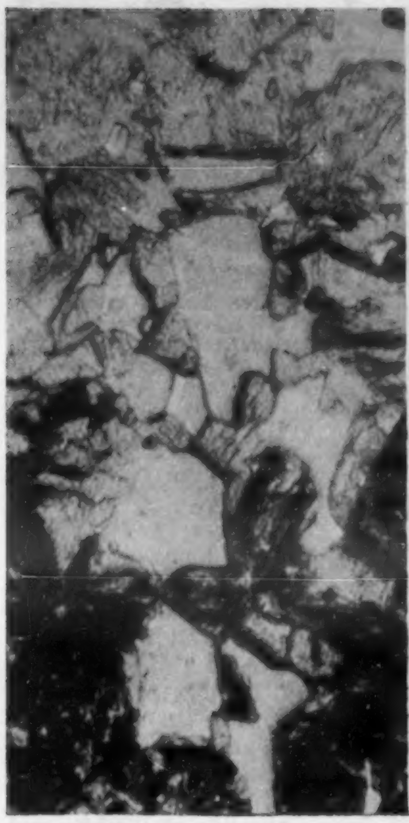

Fig. 1—.20% Carbon Steel Quenched from 1420°F. 1000X

Fig. 2—.20% Carbon Steel Quenched from 1450°F. 1000X

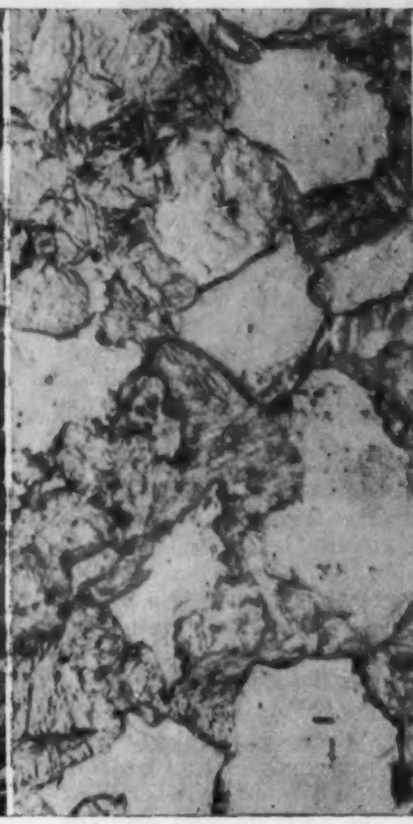

Fig. 3—.20% Carbon Steel Quenched from 1400°F. 1000X

Fig. 4—.20% Carbon Steel Quenched from 1400°F. and Tempered at 800°F. 1000X

The color of the martensitic areas in an etched specimen of quenched steel is probably largely dependent on the size of the cementite particles. In Sauveur's[6] summary of some of Lucas' work untempered martensite was found to be very lightly colored, while a decided darkening of the martensite needles was found on tempering at 400°F. The dark color of troostite is well known and is ordinarily explained as being due to the critical size of the cementite particles. In accordance with these facts, it may be assumed that up to a certain point the color of the martensitic areas is an indication of the size of the precipitated carbide particles.

Since 1340°F. was just above the Ac_1 of the steel used in this investigation, the martensitic areas retained on quenching from this temperature contained about 0.90% carbon. As the quench temperature was raised, the carbon content of the retained martensite decreased and its relative volume increased, due to the absorption of ferrite, until on quenching from Ac_3 the steel was composed entirely of a martensite or troosto-martensite containing 0.20% carbon.

As a consequence of the effect of carbon content on martensite decomposition, it might be expected that the martensite obtained on quenching from just above Ac_1 would be less decomposed than that obtained from quenching above Ac_3. Since the decomposition results in the formation and growth of minute cementite particles, the martensitic areas in a 0.20% carbon steel quenched from 1340°F. would probably be lighter in color than those in the same steel quenched from 1500°F. As previously noted and shown in Figs. 2 and 3, the martensite retained on quenching from 1400°F. and below are light tan in color, while those retained on quenching from 1450°F. and above are black. In the specimen quenched from 1420°F. the black and tan varieties are both present, as shown in Fig. 1.

These conditions lead to the conclusion that the martensitic areas in the specimens quenched from 1400°F. and below consist largely of tetragonal martensite with perhaps a very minute precipitation of carbide cementite particles. The areas in the specimens quenched from 1450°F. and above consist of a decomposed martensite in which the carbide particles have grown almost to the critical size found in troostite. This view is strongly supported by the similarity of the structures seen in Figs. 2 and 4, one of which was obtained on a quench from 1450° F. and the other on a quench from 1400°F. with a subsequent draw to 800°F. From the similarity of these two structures, as well as from the previous discussion, it is evident that the dark areas in the specimen quenched from 1450°F. (Fig. 2) are not a true martensite, but might be called troosto-martensite or secondary troostite.

On drawing specimens quenched from 1340° and 1400°F., the structures obtained are those which would be expected when any steel is drawn from martensite through troostite and sorbite to a spheroidized condition. The only structural changes during this process are the decomposition of tetragonal martensite at about 300°F. and a subsequent growth of the cementite particles. The only structural difference, then, between troosto-martensite, troostite, and sorbite is the size of the cementite particles, and it is difficult to understand how a steel could be drawn from the martensitic to the sorbitic without passing through the troostitic stage. The dark areas in Fig. 4 etch more easily than similar areas in specimens drawn at lower temperatures, and they are produced at a draw temperature in the theoretical range for troostite. It seems very likely, then, that they are for the most part troostite. It is unlikely that pearlite in low carbon steel transforms, as suggested by Crook and Taylor, to martensite at Ac_1 on heating, and from martensite to austenite to Ac_2 because martensite, a supersaturated solution, can be formed only on cooling.

From this it would seem that the specimens quenched from 1550°F. which Crook and Taylor used in their tempering tests were not, then, true martensite but either troosto-martensite or troostite, so that no marked change in structure could be expected until a draw temperature of about 900°F. was reached, except perhaps for a gradual disappearance of the acicular structure and a gradual growth of cementite particles. Careful examination of their original photomicrographs support this supposition, as the changes in structure occurring up to 900°F. are so slight as to be scarcely noticeable. It would seem justifiable to substitute, for their schedule of structural changes on drawing, a gradual and progressive growth of the cementite particles. The straight line relation obtained by them between Brinell hardness and draw temperature supports the suggested modification, since it is the relation that might be expected from the progressive variation of a single factor, that of cementite particle size, rather than the interaction of several complex factors.

SUMMARY.

The results of this investigation seem to show that the conventional theories of heat treatment apply just as well to low carbon steels as to eutectoid or hypereutectoid steels, and suggest the following explanations of the phenomena found in the heat treatment of this steel.

Due to the effect of increasing carbon content on martensite decomposition, a 0.20% carbon steel gives martensitic areas consisting to some extent of tetragonal martensite when quenched between Ac_1 and about 1420°F.; and the same steel gives areas of troosto-martensite, or possibly troostite, when quenched between 1420° and about 1650°F. Thus a steel quenched from just over Ac_3, say at 1550°F., is to some extent troostitic—perhaps entirely so—and does not have its structure changed on drawing up to about 900°F., except for gradual growth of the cementite particles and gradual elimination of the acicular structure. The martensitic areas in a steel quenched from between Ac_1 and 1420°F. will, however, draw in the usual manner through troostite and sorbite to a spheroidized condition.

[6]The Metallography and Heat Treatment of Iron and Steel. McGraw-Hill Book Co., 1912, pages 259-262.

THE ROCKWELL SUPERFICIAL HARDNESS TESTER

By V. E. LYSAGHT*

RAPID growth in the use of the nitriding process for the surface hardening of steels has created a demand from the industry for a means of measuring the hardness of this superficial layer or case. Literature on nitriding raises the questions, "How hard is the case of nitrided steel?" and "How does the hardness vary from the surface to the core?" This latter calls for a method of determining the hardness gradation or hardness depth characteristic of the nitrided case.

During the past 10 years there has been a growing tendency for 100% production testing for hardness on all heat treated parts and in the majority of heat treating plants today it is current practice to Rockwell each part after it has been heat treated. Consequently with the new art of nitriding such parts as shafts, gears, bearings, cams, chucks, gages, pins, dies, bushings, etc., which had been previously heat treated and tested, it was only logical that these parts should be tested 100% for hardness when nitrided, in order to eliminate any parts defective in hardening. In addition to 100% testing of every part it has been advocated that test pieces be nitrided in each furnace heat and that such pieces be tested for hardness depth characteristics.

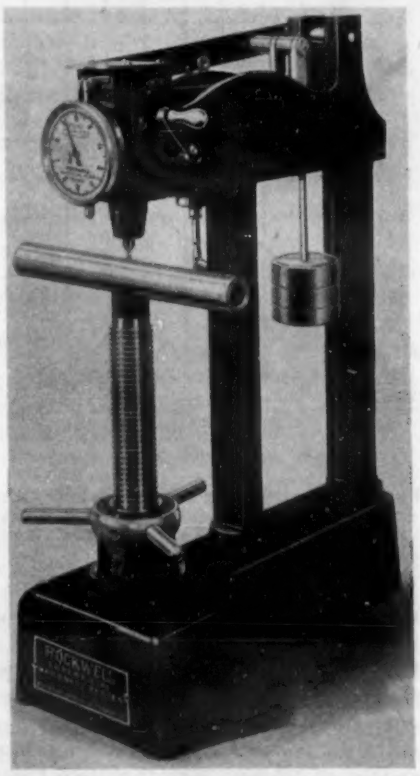

Fig. 1 — Rockwell Superficial Hardness Tester.

In order to design a machine to test the hardness of the nitrided case successfully, it was first necessary to determine just what were the requirements of the test. Early writers differed in their opinions. Some wanted a test which measured the hardness of the surface only, others a test the result of which was influenced also by the hardness beneath the surface. Some metallurgists wanted the test to determine the ductility of the case as well as the hardness; others wanted a hardness test only and had other means of examining the case for ductility.

An intensive study was made of papers on nitriding and this information, together with that derived from interviews and study in laboratories with men engaged in research and production nitriding, indicated two groups of requirements for a successful hardness test by the penetration method, the first group dealing with what will be called the penetration requirements and the second group with the mechanical requirements of the tester.

The first problem in connection with the penetration requirements was, of course, the depth of impression. To be certain, some authorities wished a test of the hardness of the surface of the case only, but the greater demand was for a test which furnished some integrated value of the spot being tested and the metal directly beneath. Such a test would credit the surface being tested with a result either harder or softer than it actually possessed, but if the depth of impression were kept small enough such an influence would be beyond the scope of testing for production and practical research purposes and only of importance when carrying on very theoretical research work.

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The second problem in connection with the penetration requirements was that of the penetrator causing the case to spall because of the case being too brittle. While it was not the purpose of this investigation to design a machine to determine the ductility of the case, it would be objectionable to spall work which was satisfactory for service, and if a penetrator could be used which spalled only work which was too brittle, it would be advantageous to the operator to detect this flaw at once, and incidentally.

The third requirement under this group related to sensitivity. The machine should be sensitive enough to pick out differences in hardness obtained by varying one or more of the controlling factors in nitriding. This sensitivity called for the greatest accuracy in a depth measuring system so that small differences in depth of impression would be transmitted to the indicating dial, yet the total depth of impression had to be shallow so that the machine readings would not be influenced by the material too far below the surface being tested.

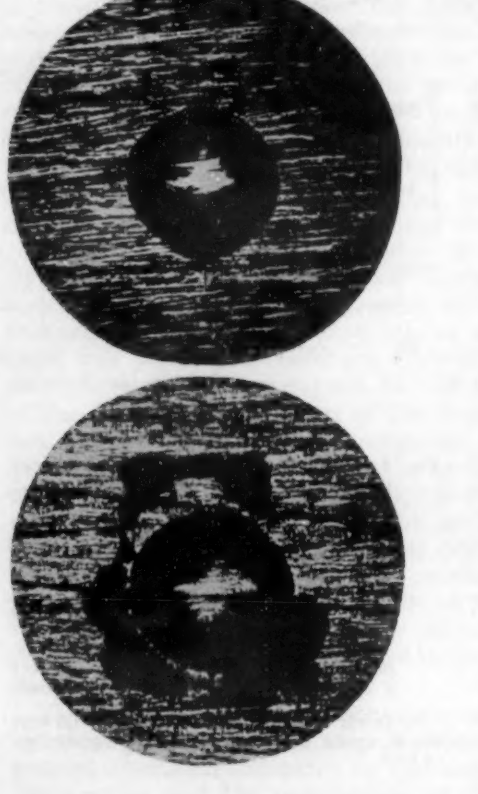

Fig. 2

The mechanical requirements for the tester were that it should be (1) rapid, (2) convenient, (3) accurate, (4) universal and (5) automatic. (1) The speed of the test should be such that a large number of tests could be made a day, but without interfering with the accuracy of the test. (2) The operation should be simple, imposing no physical fatigue or eye strain on the operator. (3) It should be possible to test odd shapes and irregular surfaces provided such pieces are properly supported against tendency to move or shift when being tested. (4) By universal is meant that the tester should be suitable for use in both the laboratory and the shop so that the metallurgist, engineer and shop man can all talk the same hardness language. This required a combination of ruggedness and precision. (5) The machine should be automatic to such an extent that the skill or technique of the operator would not influence the readings. Furthermore, the readings should be indicated directly so that the personal element would not influence the results.

G. M. Eaton[1] in one of the earliest papers on the hardness of nitrided steel, points out that using the regular Rockwell tester with 60 kg., the load is undesirably heavy, as a great range in gradations of hardness is involved in the penetrated material. In a discussion of this paper V. O. Homerberg confirms Eaton's statement. This is the same viewpoint that the manufacturers of the Rockwell tester hold, and it has never been recommended that the regular Rockwell tester be used for nitrided work except where a deep nitrided case of over 0.020" was being tested.

In order to understand why the Rockwell tester was not recommended for testing nitrided surfaces even if the major load were reduced below 60 kg., the lightest load furnished with the machine, it is necessary to study the theory of the Rockwell test.

The Rockwell hardness number is based on the difference between the depths of penetration of major and minor loads. This difference is automatically registered on an indicator dial,

the scale of which is reversed so that the harder materials will show the higher hardness numbers. In the regular Rockwell tester the minor load is 10 kg. and the major loads either 60, 100 or 150 kg. The penetrator for hardened steels is a diamond cone, the apex of which is rounded to form a spherical segment.

When testing nitrided steels the depth of impression caused by the minor load of 10 kg. is approximately 0.0002". If the major load is 60 kg. the total depth of impression is about 0.0012" and as depth of penetration is affected by the supporting material to a depth of about 10 times the depth of impression, then with a case depth of only 0.010" there is a possibility that the hardness of the core is being measured with the hardness of the case. If the major load is reduced to 30 kg. the total depth of impression is approximately .0006" and a safe factor of safety is allowed in measuring all but extremely thin cases. But the disadvantage would be that the minor load impression is still .0002" or 33⅓% of the total impression depth. Therefore, one-third of the total impression depth is not being measured and the hardness of this one-third is of vital importance and should be measured.

The next logical step in this study was to observe what the effect would be if no minor load were used. The object of the minor load is to firmly seat the penetrator in the specimen and thus hold it in position and likewise to render the test independent of the surface finish of the specimen. With no minor load it was found after a short study that the specimen required too much surface preparation to obtain reliable and consistent results. Small surface scratches or irregularities would not allow the penetrator to seat itself firmly on the surface and variations occurred in the readings which were not due to variations in hardness. Some minor load between zero and 10 kg. was manifestly needed.

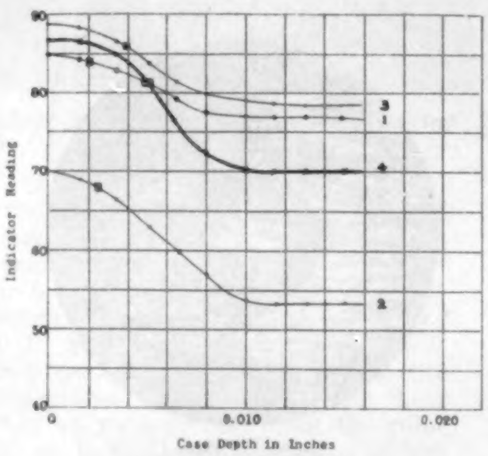

Fig. 3 — No. 1. Regular Rockwell Tester. 10 kg. minor load—60 kg. major load; standard depth measuring system.

No. 2. Same as No. 1 except using special double sensitivity depth measuring system.

No. 3. Same as No. 2 except that major load is reduced to 30 kg.

No. 4. Superficial Rockwell Tester 3 kg. minor load—30 kg. major load.

Indicates hardness reading of a C 65 test block made by each of above conditions of test.

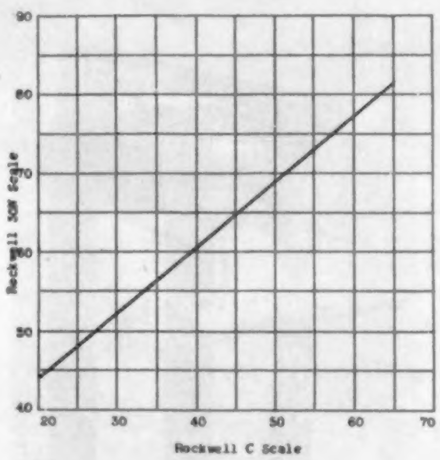

Fig. 4 — Showing the relation between the Rockwell 3ON Scale (Rockwell Superficial Hardness Tester) and C Scale (regular Rockwell Tester).

As the major load was reduced below 60 kg. the difference in hardness numbers due to change in hardness became too small and a more sensitive depth measuring system had to be used. It was necessary that errors in the depth measuring system be kept within very small limits because a depth measuring system with twice the sensitivity of the regular Rockwell but with equal tolerance of error would not be any gain.

The Rockwell Superficial Hardness Tester, Fig. 1, was designed after the above study of the problems involved in the testing of nitrided steels had been completed. In appearance the machine does not differ materially from the regular Rockwell, but the loading and depth measuring systems are radically different. The hardness number obtained in this Superficial model is based upon the regular Rockwell principle—namely an increment in depth due to an increment in load. The initial or minor load is 3 kg. and the major load either 15, 30 or 45 kg.

To compensate for loss of sensitivity caused by using lighter load, a depth measuring system was devised, twice as sensitive as the regular Rockwell, and giving (when using the light minor load found to be the most appropriate, and a major load as light as 30 kg.) a sensitivity which compares favorably with the sensitivity of the standard Rockwell "C" scale (150 kg. load and Brale penetrator). With these lighter loads it was necessary to eliminate practically all friction errors. To accomplish this, many moving parts were redesigned. It may be of interest to mention that many of these new parts are now being used in the regular Rockwell tester. The speed of application of the major load is controlled by a dashpot and valve, giving a variation of from 2 to 20 seconds.

Other mechanical features of the machine are the same as have proven to be satisfactory on the regular Rockwell tester. The machine is rapid and can be used in both shop and laboratory. Its accuracy is such that readings may be interchanged freely even between companies in different countries.

As to penetrators and the relation of their shapes to the sensitivity of the machine and the spalling of the case: The regular Rockwell penetrator is a cone with a definitely rounded apex. The advantages of this shape are that no objectionably deep penetration is caused by the minor load, that the effect of the supporting anvil is lessened when thin material is being tested, and that the rounded apex increases the life of the cone. In general the smaller the angle of a cone and the smaller the radius of a sphere, the more sensitive the penetrator becomes. But a true cone or small radius sphere, when ground from a diamond, will break sooner than a blunt point; moreover a sharp point causes too deep a penetration under the minor load.

It was desired not to use a spherical indenting tool due to the lack of geometrical similarity of impressions: As the impressions of a sphere become deeper, too great variation in hardness results from the relatively large increase in depth of impression. This is not so when using only the tip of a sphere, when variations in hardness give only small changes in depth of impression.

The next consideration was a combination of the two—a cone with a definite rounded apex, the spherical part being used on the hard surface and the sloping sides of the cone coming into the metal as the penetrator sank more deeply. Cones from 80° to 120° with spherical segments tangent to the slope of the cone with radii from 0.05 to 0.33 mm. were studied both for sensitivity and as a means of measuring the ductility of the case. Although the smaller angle cones and small radii have a little greater sensitivity on extremely hard surfaces, these points broke very quickly in production tests. The regular Rockwell "Brale" penetrator—a 120° cone with spherical tip of 0.2 mm. radius—was found to be the most satisfactory shape with regard to practicability and sensitivity. Furthermore the Brale had a tendency to spall only brittle cases, and neither spalled nor cracked any nitrided case considered suitable for service. This was when the major load was not in excess of 30 kg. Fig. 2 shows the photomicrographs of a Brale impression using a 30 kg. major load. Both specimens indicate the same hardness, yet one case spalls while the other case does not show any spalling. As this is a very hard case (30N88) and the impression is magnified 100 times, too much precision must not be expected in the contour of impression.

Because of the shallow impression employed and the sensitivity of the depth measuring system, it was necessary to grind the spherical segment of the diamond penetrator, or Brale, to conform exactly to the 0.2 mm. radius. Small differences in this radius which did not influence the reading of the penetrator when used with the regular Rockwell caused variations in readings on the Superficial model and the diamonds ground for use with this machine will be marked "N" Brales.

Pyramidal shapes of various angles and different number of sides were tested but no advantage was gained using these shapes with a depth measuring system. The spalling effect of a pyramid was found to be severe and all pyramids investigated spalled work which was apparently satisfactory for service. Eaton[2] explains this as being due to the localized plastic flow with a pyramid indenture while the flow around a conical indenture is evenly distributed. A true pyramidal penetrator would have the disadvantage of a deep minor load impression and the apex could not be tangentially rounded off into a spherical tip.

From the beginning it seemed advisable to use a new and special scale with the Superficial model of the Rockwell. It is admitted that the industry has associated certain high Brinell values with the hardness of nitrided cases but such figures are not actually obtained on the Brinell machine and much con-

fusion and misunderstanding has followed the adoption of these values. A. B. Kinzel[3] remarks that no attempt was made to convert certain case hardness values obtained by other means to Brinell scale and then explains why this was not done.

Dr. J. A. Brinell defined the indentation number which now bears his name as the ratio $\frac{P}{A}$, P being the load on the indenting tool in kilograms and A the surface area of the indentation in square millimeters. Investigation has shown that this Brinell number, as defined above, is not a constant of the test material; that is, the area of an indentation is not directly proportional to the load.[4] The theoretical discussion and causes for this lack of proportionality are too complex to be given here, but a study of the works of Meyer, Ludwik, Batson, Hoyt and others throws much light on the subject. S. N. Petrenko[5] points out that in order to secure comparable results there has been, therefore, in the past a tendency to standardize the load and the indenting tool. Thus, in the Brinell test the Brinell number is obtained with a 10 mm. ball and two loads; 3,000 kg. for hard materials, and 500 kg. for soft materials.

It is well known that when the Brinell number exceeds about 500, appreciable flattening of the ball occurs. According to G. A. Hankins[6] a Brinell hardness of 440 is sufficient to produce a marked effect on the result due to flattening of the ball. Superhard cold worked and diamond balls have been used but these do not give entire satisfaction in the measurement of hardness of hard materials.

Consideration of the above shows that unless Brinell numbers are obtained under the usual conditions of test they do not bear true relation to the ordinary conception of the Brinell number. Hardness numbers obtained using different size and shaped penetrators and varying loads, and then converted into an equivalent Brinell number simply by using the ratio $\frac{P}{A}$ should not be used.

However, even if there existed a Brinell hardness scale up to the range of 1,000 Brinell, such a scale would not be used on the Rockwell Superficial model. The principle of the Rockwell test based upon the measurement of the depth of a superimposed impression could be expressed in terms of a Brinell number only by conversion. Since any conversion is very vague and leaves so many loopholes for errors, any resultant Brinell number obtained by conversion from a Rockwell number must necessarily be indefinite and unreliable; such a means could not be used to indicate accurately small differences in depths of impression.

The following system will be used for recording readings obtained on the Superficial model of the Rockwell. The letter "N" as a prefix to the dial reading has been selected to designate readings obtained in the Superficial model with the special diamond "N" penetrator. This letter "N" is itself prefixed by major load used and followed by the hardness reading. For example, a piece of nitrided steel tested in the Superficial model, using a 30 kg. major load, diamond penetrator, and having the pointer on the dial indicate 75, would be recorded as 30N75.

Fig. 3 illustrates what has been accomplished in sensitivity and depth of impression with the Superficial model of the Rockwell. These are hardness depth characteristic curves of nitrided steel. In studying these curves bear in mind that this is a special nitrided steel with a core hardness of Rockwell C 50. Similar curves would be obtained using a Nitralloy 125 or 135 specimen. The small square (□) on each curve indicates the hardness reading obtained on a C 65 test block under the conditions of test for the particular curve. The difference in hardness between the surface of the nitrided case and a hardened steel block (C 65) can be determined by subtracting the hardness reading as shown by the small square (□) and the reading at zero depth of case. Curve No. 1 shows the results obtained on a tapered ground nitrided sample using the regular Rockwell and 60 kg. load (Special Rockwell A scale). Curve No. 2 shows the results obtained on the same sample using the regular Rockwell but with the depth measuring system the same as used with the Superficial model. The sensitivity is greatly increased but the relation of the case hardness to a C 65 test block is about the same on account of the Rockwell depth of minor and major load impressions. Curve No. 3 shows the effect of reducing the major load to 30 kg. and the sensitive measuring system on the regular Rockwell but here again the relation of the case hardness to C 65 test block is not brought out because of the deep minor load impression. Curve No. 4 shows the result obtained with the Superficial model using a major load of 30 kg. Curves No. 2 and No. 4 show about the same sensitivity but note that in curve No. 2 there are only 2 points of difference in hardness between the case hardness and a C 65 steel, while curve No. 4 shows 6 points difference.

Fig. 3 may also serve to point out that even if the major load on the Superficial machine were increased to that employed on a regular Rockwell tester, the readings would not agree with those obtained in the regular machine. The corallary of this is likewise true; if the major load in the regular Rockwell is decreased to that of the Superficial machine, Superficial hardness number will not be obtained.

With the advent of any new technical apparatus the engineer, or in this case the metallurgist and the engineer, is interested in associating the results obtained with other results with which he is already acquainted. Due to the rapid growth of the Rockwell test all parties interested in the working of metals know and understand Rockwell hardness numbers. In order to tie up the Rockwell Superficial 30N scale (the scale used for nearly all nitrided tests) with some known scale of hardness, Fig. 4 is published. The curve shows the relation between the Rockwell C scale and the 30N Rockwell Superficial scale.

It must be emphasized at this point that this conversion is to be used only to associate the relative values of the 2 scales. It does not apply on nitrided work or thin specimens where the regular Rockwell test cannot be properly used on account of the effect of the supporting anvil. The Rockwell C scale readings are obtained only by using the Brale penetrator, 10 kg. minor load and 150 kg. major load. No other condition of test will give true C scale readings. It is not within the scope of this paper to go into a detailed report on the subject of conversion. It will be sufficient to remark that conversion from one hardness to another is of approximate validity for any one grade of material, provided the material is of uniform hardness. When a hardness gradient exists, as with nitrided steel, case hardened steels, steels with decarburized surfaces, etc., conversion cannot be resorted to under any conditions because of different penetration depth. The data for the curve shown in Fig. 3 were obtained on uniformly hardened and tempered steel blocks 1/4" thick and the average of several tests plotted.

The question naturally arises as to the possibility of using the Rockwell Superficial hardness tester on work which was previously tested on the regular Rockwell. Because of the light minor load used with the Superficial model, the surface condition of the steel will greatly influence the results. Slight surface decarburization, tool marks and surface finish of the steel will affect the readings. This is as it should be, for the Superficial model is designed to test the surface conditions.

The depth of impression of the regular Rockwell penetrator has been found to give the engineer and the metallurgist the desired information regarding the hardness of any homogeneous metal. Any attempt to use a more shallow impression will cause confusion and unless the steel or metal is uniformly hardened and free from any purely surface condition, no attempt should be made to use the Superficial model, unless it is desired to study this surface condition. The depth of penetration of the regular Rockwell was carefully selected and that it gave the desired result is manifest by the large number of machines now in use.

The depth hardness characteristic of a nitrided case obtained by taking hardness readings at definite spots on a tapered ground specimen indicates to the metallurgist the hardness of the surface and the penetration of the hardness. That such a characteristic is the information that interests the metallurgist is shown by a survey of the literature accumulating on the nitriding process. Nearly every paper on nitriding contains one or more depth-hardness characteristic curve. Figs. 5 and 6 show typical depth hardness characteristics for different samples. The samples were tested using the diamond Brale penetrator with 30 kg. major load on the Superficial model of the Rockwell tester. Except when the nitrided case is extremely shallow or very deep the 30 kg. load gives the best results. The depth of penetration into the nitrided case is only about 0.00065". One division of penetration of the Superficial model represents a depth of 0.00004". By subtracting the dial reading from 100 the depth of penetration can be readily determined by multiplying this difference by 0.00004". This depth will be in error by the depth of the minor load impression. However, with nitrided and hardened surface this minor load depth is small, being in the neighborhood of 0.00005".

Fig. 5 shows the depth hardness characteristics obtained on tapered ground samples which were nitrided at 975° F. for 12, 24 and 72 hours. The samples were ground at a 1½° angle and readings taken every 1/16" and plotted to the nearest 0.0005". No difference in readings resulted from the slight angle at which the samples were tapered. The surface readings were taken after the surface was polished with fine emery paper.

All the results shown in Figs. 5, 6 and 7 were obtained on Nitralloy 125. The steels were nitrided by the Leeds & Northrup Company in an early model of Homo nitriding furnace and represent work which was completed 2 years ago. Since then

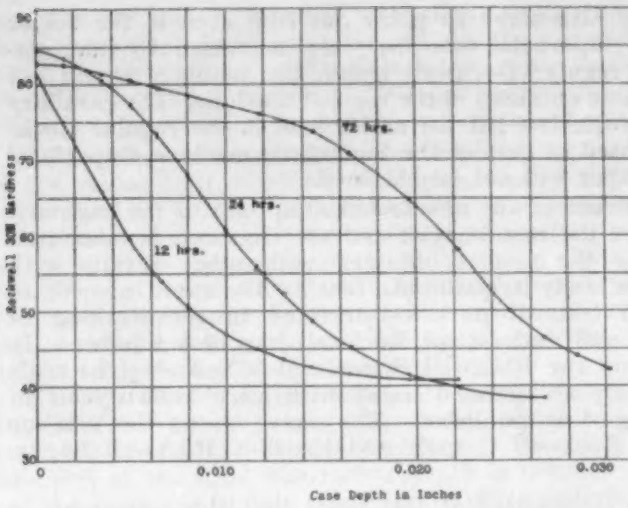

Fig. 5 — Showing the depth hardness characteristics of steel nitrided at constant temperature for 12, 24 and 72 hours.

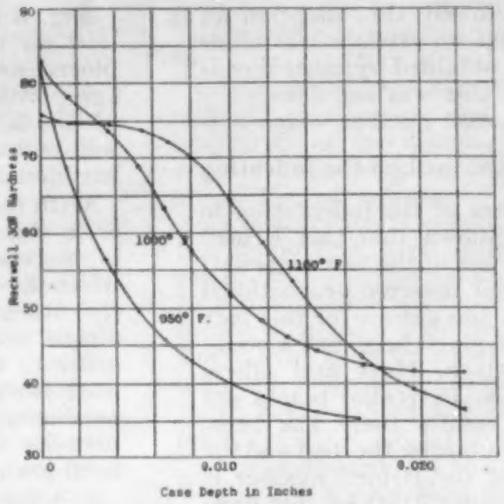

Fig. 6 — Showing the depth hardness characteristic for steel nitrided for the same time but with the temperature varied from 950° to 1100°F.

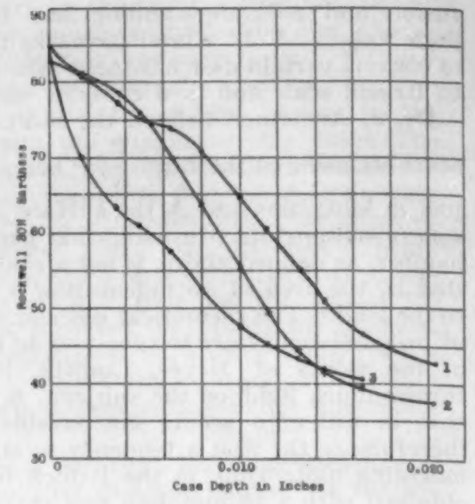

Fig. 7 — Showing the depth hardness characteristics of steel nitrided using a progressive temperature cycle.

rapid progress has been made both in the alloys to be nitrided and in the design of the furnace, with the result that both harder cases and deeper penetration can now be obtained. This should be considered in studying the results. However, the curves do bring out the sensitivity and characteristics of the machine and as it is not the purpose of this paper to study the nitriding process, the work was not repeated on samples nitrided with the latest equipment.

Fig. 6 shows the results obtained on specimens nitrided for 24 hours with the temperature varied from 950° to 1100° F.

Fig. 7 shows the results obtained on specimens nitrided using a progressive temperature cycle. The time and temperatures are marked on the curve.

Table 1 shows the readings obtained on 20 samples of airplane parts which were nitrided under commercial production conditions. The readings represent the average of 3 tests on each surface and the average variation over the surface was about one point. This table clearly brings out the need for production testing.

Table 1

Sample Number	Rockwell 30N Reading	Sample Number	Rockwell 30N Reading
1	30N 83.0	11	76.0
2	82.5	12	77.0
3	82.5	13	84.0
4	82.5	14	82.5
5	74.5	15	75.0
6	74.0	16	80.0
7	85.0	17	85.0
8	82.5	18	77.5
9	73.0	19	79.5
10	73.0	20	76.0

Because of the shallow impressions employed by the Rockwell Superficial hardness tester, it can be used for testing very thin material such as hardened strip steel, annealed steel, brass, etc. Thin razor blade steel (.006" in thickness) is being tested successfully by this model Rockwell.

When testing soft material the 1/16" steel ball penetrator is used. The problem of testing thin material is one which is closely related to measuring the hardness of a thin superficial surface and the results obtained with the Rockwell Superficial hardness tester on this material are now being studied.

The author wishes to acknowledge the assistance received from H. W. McQuaid of the Timken-Detroit Axle Company, W. Richison Schofield and J. W. Harsch of the Leeds & Northrup Company and Dr. V. O. Homerberg and Dr. J. P. Walsted of the Massachusetts Institute of Technology, all of whom allowed the use of their laboratories and samples for the investigation.

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[3]*Special Edition Transactions American Society for Steel Treating*, Oct. 1929, page 182.

[4]*Bureau of Standards, Journal of Research*, Vol. 5, July 1930, pages 19-50. *Research Paper* 185.

[5]*Bureau of Standards Technologic Paper* No. 334.

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Copper-Tin-Lead System

(Continued from Page 182)

almost the same range of composition as those long used in practice (Fig. 7).†

However, the development of Cu-Sn-Pb alloys is not yet finished, for the properties of the alloys which are heterogeneous in the liquid state and which it is possible in principle to produce have not been investigated. The use of annealed or wrought materials is also conceivable, if the change of the mechanical properties produced by annealing is compensated for by a higher Sn or Zn content, while the high Pb content can be retained without regard to the course of the limit of the miscibility gap in the liquid state.

†Editor's note: Mr. E. R. Darby of the Federal-Mogul Corporation states, "It would appear that all of the usual bearing compositions are within the area of complete miscibility. This is in accord with our own experiments.

"Two years ago we attempted to explain some of the peculiar actions of the alloy 70-5-25 on the thought that we might be dealing with liquid immiscibility. Samples were taken for analysis from the bottom, middle and top of a 200 lb. ladle by tapping from holes provided in the side of the ladle plugged with clay. The metal was allowed to stand for ten minutes before sampling.

"No marked difference in lead content was found. This was the case with all our alloys except one where 0.25% phosphorous was present with 10% tin and 20% lead. Here we found a decided falling off in lead from the upper levels."

Dr. G. K. Burgess

Dr. G. K. Burgess, Director of the Bureau of Standards, died of cerebral hemorrhage on July 2. Trained as a physicist, Dr. Burgess, after a brief teaching career, entered the Bureau as one of its first employees. In his work in the Heat Division, he became interested in metallurgical problems and personally carried out many fundamental investigations of high precision. As the work of the Bureau grew, Dr. Burgess became impressed with the importance of metallurgical research and brought about the formation of the Division of Metallurgy, of which he became chief, at its establishment. His reputation in metallurgy brought him recognition in the form of the presidencies of the A.S.S.T. and the A.S.T.M., life membership in the A.F.A., and many responsible committee obligations in these and other metallurgical societies. During and after the war, he served on many government boards. His appointment to the Directorship of the Bureau took him to a larger field of activity and forced a diminution in his active metallurgical contacts. While most obituary notices refer to him as a physicist and an administrator, he was a metallurgist through a large part of his career. Metallurgy has lost one of its finest guiding spirits.

Correction: Copper Tin Compound in Babbitt

The illustration marked Fig. 7 opposite page 170, Metals & Alloys, July issue should be Fig. 8. The illustration marked Fig. 8 should be Fig. 7.